JOURNAL of THE IRON AND STEEL INSTITUTE

VOL. 173 PART 4

APRIL 1953

PUBLISHED MONTHLY

Editorial contents copyright by
THE IRON AND STEEL INSTITUTE

EDITOR
K. Headlam-Morley

MANAGING EDITOR
E. R. Mort, B.Sc., F.I.M.

ASSISTANT EDITOR
C. R. Harman, B.Sc.(Eng.)

EDITORIAL AND ADVERTISEMENT OFFICES 4 Grosvenor Gardens, London, S.W.1 · SLOane 0061
Telegrams · · Irosamente, Sowest

SUBSCRIPTIONS
Members Free : Extra Copies 7s. 6d. each
Non-Members : £5. 0. 0 per annum
Single Copies 10s. each

NOTE
The Institute as a body is not responsible for statements made, or opinions expressed, in papers, articles, or discussion appearing in the Journal of The Iron and Steel Institute.

CONTENTS

PROCEEDINGS OF THE INSTITUTE

Report of Council for 1952 329

IRON AND STEEL INSTITUTE PAPERS

Carbides in Chromium, Molybdenum, and Tungsten Steels by Kehsin Kuo 363

The Deformation of Austenite in Relation to the Hardness Characteristics of Steel by G. R. Bish and H. O'Neill 398

Effect of Boron on the Mechanical Properties of Low-Alloy Steels by R. Wilcock 406

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Control of Strip Thickness in Cold Rolling by Varying the Applied Tensions by R. B. Sims, J. A. Place, and P. R. A. Briggs 343

Works Trial of the 'T' Method of Automatic Gauge Control : A Progress Report by R. B. Sims, J. A. Place, and P. R. A. Briggs 354

Narrow-Side Vertical Cracking in Ten-Ton Ingot Moulds by A. Jackson and A. N. Whiting 360

Effect of Arsenic and Antimony on Temper-Brittleness by G. W. Austin, A. R. Entwisle, and G. C. Smith 376

Temper-Brittleness in High-Purity Iron-Base Alloys by A. Preece and R. D. Carter 387

IRON AND STEEL ENGINEERS GROUP

Discussion on Electric Drives for Continuous Hot Mills 420

Discussion on a New Medium-Wide Hot Strip Mill 425

NEWS

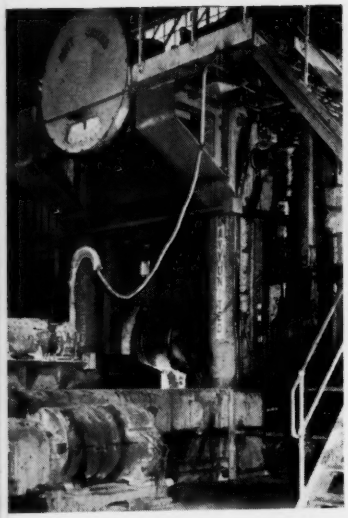
The Iron and Steel Institute ... 431
News of Members 432
Contributors to the Journal ... 432
Iron and Steel Engineers Group 433
British Iron and Steel Research Association 433
Institution of Metallurgists ... 433
News of Science and Industry 434
Diary 434
Translation Service 434

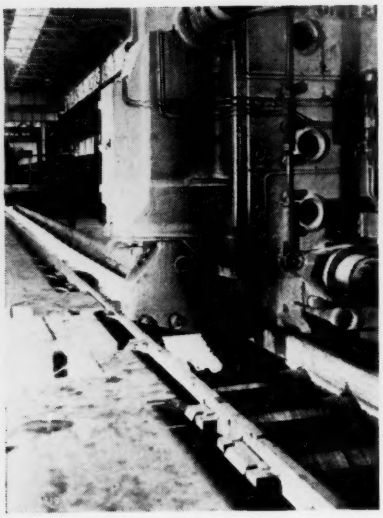
ABSTRACTS

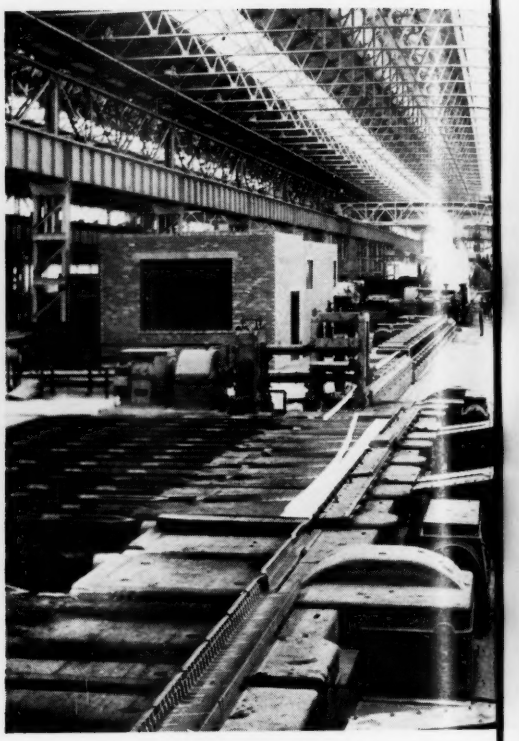
Contents 435
Abstracts 435-470
Book Notices 470
New Publications 472

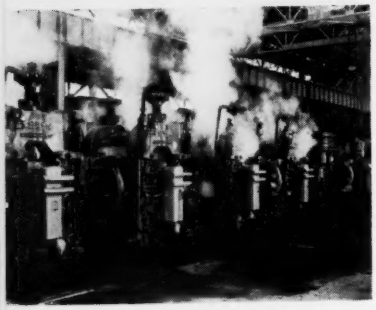
ONE OF THE IMPORTANT FEATURES OF THIS MILL IS ITS

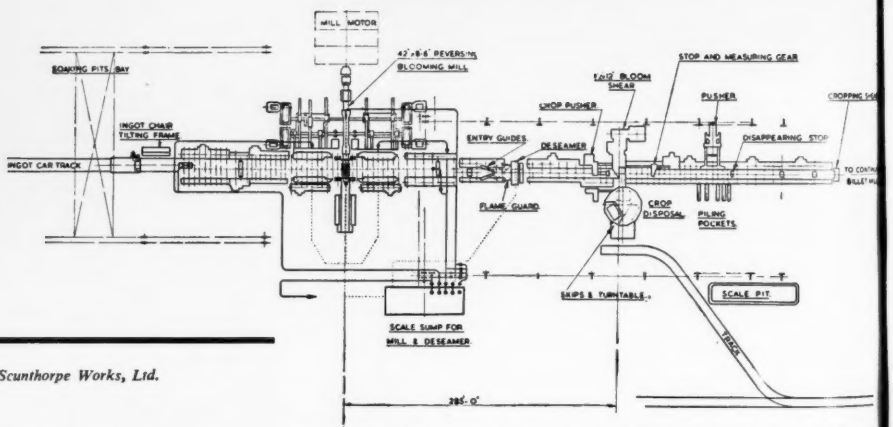
planned layout











LAYOUT OF 42" x 8' 6" REVERSING BLOOMING MILL AND BLOOM SH

Illustrations by courtesy of JOHN LYSAGHT'S Scunthorpe Works, Ltd.

A "greenfield" site gives a mill designer a rare opportunity. It allows him to put forward a layout hand-tailored to suit the productive requirements set and planned from start to finish to give the most efficient and economic flow of material.

The plant illustrated was designed and laid out specifically to produce upwards of 500,000 tons of billets annually. The installation comprises blooming mill, deseamer, bloom shears, six-stand billet mill, flying shears and cooling banks. Employing straight-line delivery from soaking pits onwards it enables high rate billet production without the necessity for reheating, offers unimpeded access for maintenance, makes adequate provision for future extensions and generally represents capital employed to profitable advantage, the acid test of any mill layout.

DAVY-UNITED

SHEFFIELD · MIDDLESBROUGH · GLASGOW

progressive engineering at the service of the metals industries

RICHARD MATHER, B.Met.(Hons.), F.I.M.

RICHARD MATHER, Chairman and Managing Director of the Skinningrove Iron Company, was born in 1886. He was educated at the Royal Grammar School, Sheffield, and later at the University of Sheffield, where he obtained an honours degree in metallurgy in 1907. After graduation, he took his first appointment as a metallurgist with the firm of Cochrane and Company, Ironfounders, of Middlesbrough. Four years later he joined the staff of the Research Department, Woolwich, also as a metallurgist, and remained there until 1919. From that time until 1925, Mr. Mather was Metallurgical Inspector to the Government of India, and during two periods, from 1923 to 1924 and from 1926 to 1927, he was Technical Adviser to the Indian Tariff Board.

Mr. Mather re-entered the productive side of the industry in 1927, when he became Deputy General Manager of the Tata Iron and Steel Company at Jamshedpur, India. He was appointed Technical Director of that Company in 1930, and during the following ten years he was mainly responsible for plant developments which more than doubled the output of the Jamshedpur works. He returned to England in 1940, and in 1941 became Consultant to Messrs. Pease and Partners, Limited. He was appointed a Director of the Company in 1942, and in the same year he was elected Chairman and Managing Director of the Skinningrove Iron Company.

From 1946 to 1949, Mr. Mather was a member of the Iron and Steel Board of the Ministry of Supply. He was the representative of British employers at meetings of the Iron and Steel Committee of the International Labour Organisation in the U.S.A. (1946), in Sweden (1947), and at Geneva (1949); he was elected Chairman of the Employers' Group at the 1947 and 1949 meetings.

Mr. Mather became a Member of Council and of the Executive Committee of the British Iron and Steel Federation on their reconstitution towards the end of the war; he is also Chairman of two committees and a member of some others. When the British Iron and Steel Research Association was formed in 1945 he became a Member of Council, and in 1950 he was elected Chairman. He was Chairman of the North-East Coast Steelmakers' Association in 1948 and in 1952. Mr. Mather joined The Iron and Steel Institute in 1913, was elected a Vice-President in 1945, and became President in 1951. He is a Founder Fellow of the Institution of Metallurgists.

At the Annual General Meeting in 1953 Mr. Mather is to be awarded the Bessemer Medal for his distinguished services to The Iron and Steel Institute and to the industry.

Richard Mather, B.Met.(Hons.), F.I.M.
Bessemer Medallist 1953

PROCEEDINGS OF THE INSTITUTE

REPORTS OF COUNCIL - OFFICIAL NOTICES

REPORT OF COUNCIL FOR 1952

PATRON

HER Majesty Queen Elizabeth II has been graciously pleased to become Patron of the Institute.

By order of the Council, the following letter was addressed to Sir Ulick Alexander, G.C.V.O., K.C.B., C.M.G., O.B.E., Keeper of the Privy Purse.

5th June, 1952

Sir,

We are desired by the Council of The Iron and Steel Institute to ask you to submit to Her Majesty Queen Elizabeth II a humble and loyal request that Her Majesty will be graciously pleased to honour The Iron and Steel Institute by consenting to become its Patron.

The Council of The Iron and Steel Institute recall with pride that Her Majesty's royal father, His Majesty King George VI, was graciously pleased to honour the Institute by becoming its Patron on his accession; and that his predecessors on the Throne, His Majesty King Edward VII, His Majesty King George V, and His Majesty King Edward VIII, were also pleased to honour this Institute by granting their Patronage.

In transmitting this request, we are desired to ask you to assure Her Majesty of the loyalty and devotion of the Council and Members of this Institute.

We have the honour to remain,

Yours faithfully,

(*Signed*) H. LEIGHTON DAVIES, *President*

(*Signed*) K. HEADLAM-MORLEY, *Secretary*

The following reply was received:

Privy Purse Office,
Buckingham Palace
22nd July, 1952

Dear Sir,

I am commanded by The Queen to inform you that Her Majesty has been graciously pleased to grant her Patronage to The Iron and Steel Institute.

It will be in order for the words "Patron—Her Majesty The Queen" to appear in future under the name of your institute on all correspondence.

Yours truly,

(*Signed*) ULICK ALEXANDER,
Keeper of the Privy Purse

The President,
The Iron and Steel Institute.

Annual Report

The Council submit this, their Annual Report and Statement of Accounts for the year 1952, to Members for their approval at the Eighty-fourth Annual General Meeting of The Iron and Steel Institute.

The Annual and Autumn General Meetings were held in London in April and November and a successful Special Meeting was organized in South Wales during October. A joint meeting with the British Iron and Steel Research Association on the Corrosion of Steel under Phosphate Coatings was held on 28th October.

The Engineers Group organized a number of interesting meetings which were well attended.

ROLL OF THE INSTITUTE

The total membership at 31st December, 1952, was 4801, compared with 4776 at 31st December, 1951.

With the ending of the state of war with Japan on 28th April, 1952, the Council decided that those of Japanese nationality are again eligible for Membership.

During the year, 292 Ordinary and 74 Associate Members were newly elected or reinstated. The deaths of 43 were reported; 207 resigned, 76 were struck off the Roll of Members, and 15 elections lapsed for non-payment of the subscriptions due.

An analysis of the membership at 31st December, 1952, is given in Table I, whilst the membership trend since the Institute was founded is shown in Fig. 1.

OBITUARY

The Council regret to announce the deaths of Monsieur Aloyse Meyer, Président du Conseil de l'Administration de la Société Anonyme des A.R.B.E.D., Luxembourg, and of the Rt. Hon. Sir Andrew Duncan, P.C., G.B.E. M. Meyer, who joined the Institute in 1921, was elected an Honorary Vice-President in 1934 and was awarded the Bessemer Medal in 1937. Sir Andrew Duncan, who joined the Institute in 1936, was Independent chairman of the British Iron and Steel Federation from 1935 to 1940, and from 1945 until his death.

The Council also regret to announce the deaths of 43 Members; of these, the following 30 occurred during 1952:

BANKS, J. T. (Manchester)	June
BOWDEN, J. S. (Greenford, Middx.)	28th November
BRADBURY, H. T. (Sheffield)	30th January
BROOKE, W. J. (Sheffield)	24th February
CRAVATTE, ERNEST (Buenos Aires)	October
DEVEREUX, W. G. (Farnham, Bucks)	21st June
DOBSON, W. E. (London)	22nd August
DRANE, H. D. H. (Wigan)	5th April
DUNCAN, Rt. Hon. Sir ANDREW R., P.C., G.B.E. (London)	30th March
GELLNER, O. H. (London)	January
GRIFFITHS, Sir WILLIAM T.	30th July
GRIST, C. J. (Brewood, Staffs.)	February
HALET, EMILE A. (London)	24th June
HEUSSER, M. H. (Switzerland)	13th January
LESSNER, CHARLES (Spain)	Unknown
LEVICK, H. S. (Sheffield)	4th November
MEYER, A. (Luxembourg)	3rd May
MUNNICK, E. H. (S. Rhodesia)	22nd May
REASON, F. E. (Cardiff)	8th March
RIDGE-BEEDLE, P. D. (Glasgow)	8th May
SAXTON, CLEMENT (Northwood, Middx.)	June
SCOTT, W. W. (St. Louis, U.S.A.)	15th April
SHAW, H. A. (Middlesbrough)	3rd January
SILLAVAN, JAMES (Manchester)	8th June
SYKES, EDWIN (Huddersfield)	13th October
VANZETTI, Dr.Ing. G. (Milan)	5th March
VICAIRE, M. J. A. (Paris)	28th January

Table I—MEMBERSHIP OF THE IRON AND STEEL INSTITUTE AT 31ST DECEMBER, 1952, AND COMPARISON WITH 1951

	Membership at 31st December, 1952												1951		
	Ordinary Membership		Joint Membership with—						Totals		Total Membership		Grand Totals	Total Membership		Grand Totals
			Institute of Metals		Institution of Metallurgists		Both Institute and Institution									
	Home	Over-seas	Home	Over-seas	Home	Over-seas	Home	Over-seas	Home	Over-seas	Home	Over-seas		Home	Over-seas	
Patron	1	...	...	...	...	...	...	...	...	...	1	...	1	1	...	1
Members:																
Hon. Vice-Presidents	11	5	2	...	...	...	...	...	2	...	13	5	18	13	6	19
Hon. Members	3	4	1	...	...	...	...	...	1	...	4	4	8	3	4	7
Non-Paying Members	4	1	...	...	...	...	...	...	...	...	4	1	5	5	1	6
Life Members	55	22	...	3	7	...	4	...	13	3	68	25	93	73	26	99
Ordinary Members	1739	786	473	335	460	44	420	52	1353	431	3092	1217	4309	2958	1164	4122
Total (Members)	1813	818	478	338	467	44	424	52	1369	434	3182	1252	4434	3053	1201	4254
Associate Members	98	22	179	36	16	1	14	1	209	38	307	60	367	464	58	522
Total (both classes)	1911	840	657	374	483	45	438	53	1578	472	3489	1312	4801	3517	1259	4776

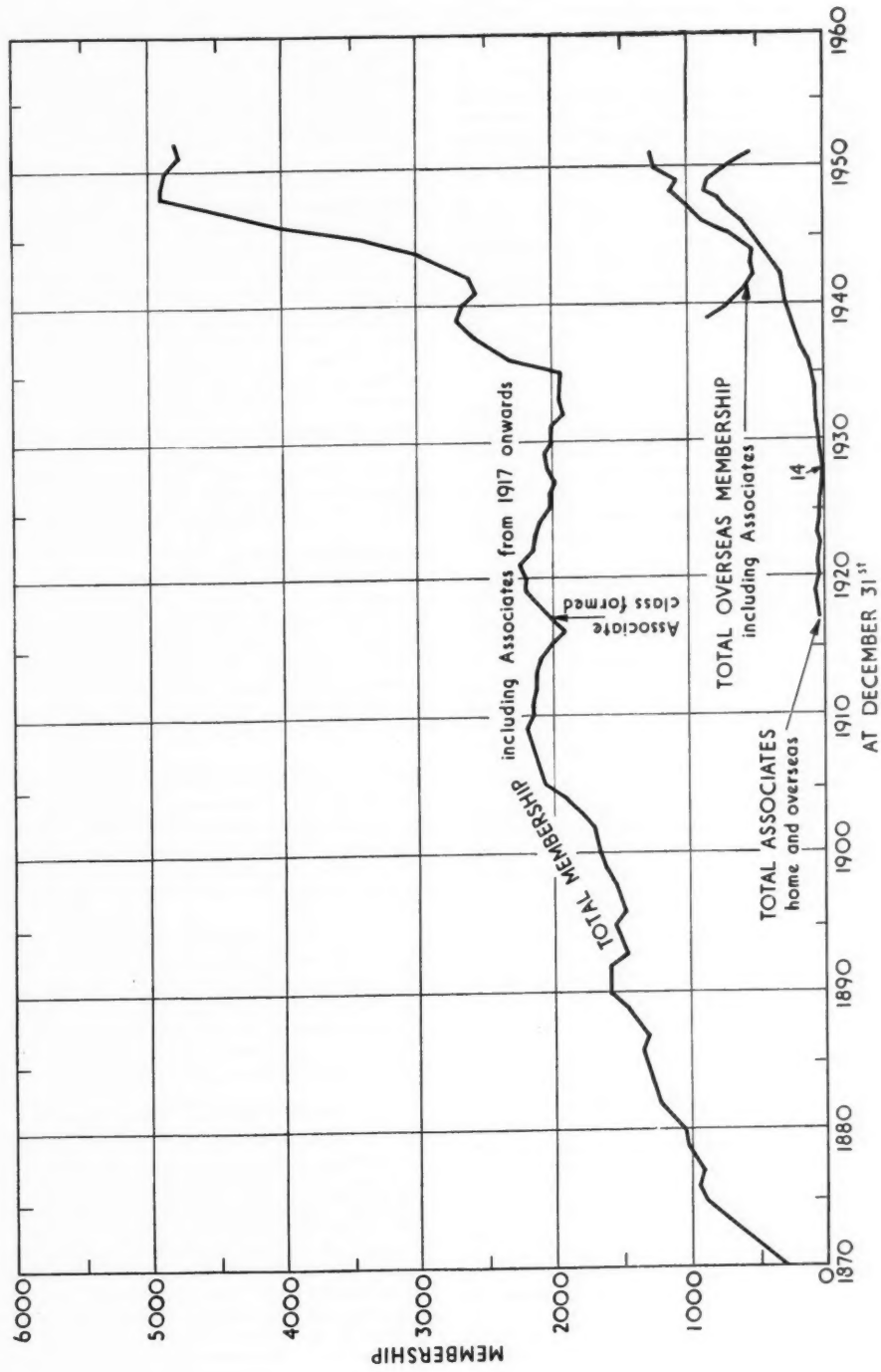


Fig. 1—Variations in Membership since the foundation of the Institute in 1869

WATERHOUSE, Dr. G. B. (Massachusetts, U.S.A.)	10th May
WELLER, F. (Sudborough, Northants)	25th May
WILLIAMS, EMRYS (Gowerton, Glam.)	23rd June

CHANGES ON THE COUNCIL

President-Elect

At the Autumn General Meeting it was announced that the Council had unanimously decided to nominate Mr. James Mitchell, C.B.E., to become President at the Annual General Meeting in 1953.

Honorary Treasurer

It was also announced that Mr. W. Barr (Member of Council) had kindly agreed to serve as Honorary Treasurer as from 1st January, 1953. The President expressed the great appreciation of the Council and of the Members of the work which Mr. James Mitchell had done as Honorary Treasurer since he had assumed the appointment on 1st January, 1949.

Other Changes

During 1952, Mr. W. B. Baxter, C.B.E. (Vice-President) was elected Hon. Vice-President, and Sir Charles Bruce-Gardner, Bt., and Mr. C. R. Wheeler, C.B.E. (Members of Council) were nominated to be Vice-Presidents. Mr. E. Julian Pode (Steel Company of Wales Ltd.) and Mr. B. Chetwynd Talbot (Cargo Fleet Iron Co., Ltd.) accepted invitations to become Members of Council.

The following agreed to serve as Honorary Members of Council during their presidencies of the Societies named: Dr. C. J. Smithells, M.C. (Institute of Metals, following Professor A. J. Murphy); Mr. F. Kennedy (Cleveland Institution of Engineers, following Mr. N. C. Lake); Mr. W. D. Jenkins (Ebbw Vale Metallurgical Society, following Mr. F. E. Probyn); Dr. N. J. Petch (Liverpool Metallurgical Society, following Mr. F. K. Neath); Mr. W. L. James (Lincolnshire Iron and Steel Institute, following Mr. A. Jackson); Mr. R. S. Brown, M.B.E. (Liverpool Metallurgical Society, following Mr. V. L. Farthing); Mr. H. Allison (Manchester Metallurgical Society, following Mr. A. B. Ashton); Mr. E. W. Colbeck (Sheffield Metallurgical Association, following Mr. E. T. Gill); Mr. G. Baker, O.B.E. (Sheffield Society of Engineers and Metallurgists, following Professor H. W. Swift); Mr. W. Regan (Staffordshire Iron and Steel Institute, following Mr. K. H. Wright); Mr. H. N. Bowen (Swansea and District Metallurgical Society, following Mr. L. A. S. Perrett). Professor R. Hay continued to serve as an Honorary Member of Council during his continued presidency of the West of Scotland Iron and Steel Institute.

Mr. E. Taylor-Austin, President of the North Wales Metallurgical Society, joined the Council as an Honorary Member when that Society became affiliated to the Institute on 1st January, 1952.

Mr. T. G. Grey-Davies (Chairman, Newport and District Metallurgical Society) accepted an invitation to become an Honorary Member of Council. Mr. G. H. Latham (Vice-President), who had been President of the Newport and District Metallurgical Society since its foundation and had represented that society on the Council, agreed that in future the Chairman of the Society should take his place as Honorary Member of Council.

Mr. R. S. Brown, M.B.E., and Dr. L. B. Pfeil, O.B.E., were appointed to represent the Institution of Metallurgists in place of Mr. W. E. Bardgett and Mr. W. W. Stevenson.

In accordance with Bye-Law No. 10, the following names of Vice-Presidents and Members of Council due to retire at the Annual General Meeting, 1953, were announced at the Autumn Meeting, 1952:

Vice-Presidents—Mr. G. Steel, The Hon. R. G. Lyttelton, and Mr. I. F. L. Elliot.

Members of Council—Mr. D. F. Campbell, Mr. G. H. Johnson, Mr. D. A. Oliver, Dr. C. Sykes, F.R.S., and Mr. W. Barr.

No other Members having been nominated up to one month before the Annual General Meeting, the retiring Members are presented for re-election.

AWARDS AND PRESENTATIONS

The Bessemer Gold Medal for 1952 was awarded to Mr. H. H. Burton, C.B.E. (English Steel Corporation, Ltd.) for his distinguished services to the steel industry and to metallurgy, with particular reference to the development of alloy steels and of heavy forgings.

The Sir Robert Hadfield Medal for 1952 was awarded to Dr. L. Reeve (Appleby-Frodingham Steel Company) in recognition of his contributions to the science and practice of ferrous metallurgy and, in particular, his researches on the weldability of low-alloy steels.

The Williams Prize for 1952 was awarded jointly to Mr. J. A. Bond (Appleby-Frodingham Steel Company) and Mr. T. Sanderson (Workington Iron and Steel Company) for their paper on "*Full-Scale Blast-Furnace Trials*" (*Journal*, 1951, vol. 168, May, pp. 24–39).

Presentation to Mr. Elsdon

At the Annual General Meeting the President made a presentation to Mr. Richard Elsdon, a member of the staff of the Institute since 1903 and Librarian since 1908, of a cheque subscribed by the Members on his retirement in 1951. He also announced that the Council had nominated Mr. Elsdon to be an Honorary Member of the Institute and proposed to arrange for a portrait of him to be painted and hung in the Library.

ANDREW CARNEGIE RESEARCH SCHOLARSHIPS

The following awards were made from the Andrew Carnegie Research Fund during 1952:

B. CINA (Sheffield University): £150 (second grant) for work on the immediate effect of cold-work on the gamma–alpha transformation characteristics of the Fe–Ni–Cr alloys.

S. GARBER (Sheffield University): £250 (second grant) for work on graphite formation in cast irons.

A. SHELTON (Imperial College of Science and Technology, London): £250 (second grant) for research on the measurement of Poisson's ratio in the elastic range of various metallic materials, and also in the plastic state at various degrees of work-hardening.

MOND NICKEL FELLOWSHIPS

Mr. F. H. Saniter continued to represent the Institute on the Mond Nickel Fellowships Committee. Mr. E. H. Jones was appointed by the Institute of Metals as their representative in succession to Mr. W. A. C. Newman.

The Committee made the following awards for 1952:

A. C. DUCE (Jos. Lucas (Gas Turbine Equipment) Ltd.): To study, in the United Kingdom, on the Continent, in the United States of America, and in Canada, the metallurgy and testing of materials, especially in sheet form, developed for high-temperature service, and the techniques employed in the manufacture of the combustion systems of gas-turbine engines.

F. G. HORTON (National Foundry College): To study casting production methods in the United Kingdom

and on the Continent, with special reference to shell moulding, centrifugal casting, and sand-cement moulding.

P. KEMPSON (Henry Wiggin and Co., Ltd.): To study the development, application, and control of ferrous and non-ferrous melting and ingot casting processes in the United Kingdom, on the Continent, in the United States of America, and in Canada.

H. A. LONGDEN (Steel, Peech and Tozer): To study metallurgical control methods in the United Kingdom, on the Continent, in the United States of America, and in Canada, with special reference to open-hearth slag and temperature control.

The Council agreed, during 1952, to a recommendation from the Mond Nickel Fellowships Committee that the Fellowship Regulations should be amended so as to raise the value of each Fellowship to between £900 and £1200, instead of between £650 and £850 as previously.

MEETINGS

Annual General Meeting and the Sixth Hatfield Memorial Lecture

The 83rd Annual General Meeting took place on 30th April and 1st May, 1952. The morning session on 30th April was held at the Royal Institution, Albemarle Street, London, W.1. Mr. R. Mather, the retiring President, inducted the President-Elect, Captain H. Leighton Davies, C.B.E., into the Chair; the new President then delivered his Presidential Address on "*The Development of the Tin plate Trade: With Particular Reference to the Installation of the Cold-Reduction Plant at Trostre, Llanelly*" (*Journal*, 1952, vol. 171, May, pp. 15–37). After a short break, Professor E. N. da C. Andrade, F.R.S., delivered the Sixth Hatfield Memorial Lecture on "*The Flow of Metals*" (*Journal*, 1952, vol. 171, July, pp. 217–228). A buffet luncheon was served in the Joint Library at 4 Grosvenor Gardens, London, S.W.1. The afternoon session on 30th April and the morning and afternoon sessions on 1st May took place at the Offices of the Institute. A Members' Dinner was held at Grosvenor House, Park Lane, London, W.1, on 29th April. The proceedings of the Meeting are recorded on pp. 25–51 of the *Journal* for September, 1952 (vol. 172).

Autumn Meeting

The Autumn General Meeting was held at the Offices of the Institute on 26th and 27th November, 1952. The President, Captain H. Leighton Davies, C.B.E., was in the Chair at the opening of the meeting and on the second day; during his absence the Chair was taken by Mr. James Mitchell, C.B.E. (Hon. Treasurer). The proceedings of the Meeting are recorded on pp. 256–258 of the *Journal* for March, 1953 (vol. 173).

Special Meeting in Swansea

A Special Meeting was held in Swansea from 7th to 9th October, 1952, by invitation of the Mayor of Swansea.

A Reception Committee was formed under the Chairmanship of Colonel J. M. Bevan, M.C., D.L., J.P., to make the arrangements in South Wales. A Ladies' Committee under the Chairmanship of Mrs. J. M. Bevan was also formed. Sir Lewis Jones, J.P., South Wales Siemens Steel Association, and Mr. P. O. Williams, Welsh Plate and Sheet Manufacturers' Association, kindly consented to act as Honorary Secretaries.

The Chairman and Board of Directors of the Steel Company of Wales Ltd. issued an open invitation to Members of the Institute and their Ladies to visit the Trostre Works of the Tinplate Division of the Company on Tuesday, 7th October. Members and Ladies were able to take part in this visit whether they took part in the whole meeting or not. About 1300 people were present. The chief guest was the Minister of Supply, the Rt. Hon. Duncan Sandys.

The President and Council of The Iron and Steel Institute wish to express their sincere thanks to the Mayor of Swansea and the Chairman and Members of the Reception and Ladies' Committees; to the Chairman and Board of Directors of the Steel Company of Wales Ltd., the South Wales Siemens Steel Association, and the Welsh Plate and Sheet Manufacturers' Association; to the Directors of the Works who kindly invited Members to visit them; and to all others who so helpfully collaborated in organizing the Meeting.

A short account of the Meeting was published on pp. 129–140 of the *Journal* for February, 1953 (vol. 173).

Meeting on Corrosion of Steel

A Meeting was held in conjunction with the British Iron and Steel Research Association, at the Offices of The Iron and Steel Institute, on Tuesday 28th October, 1952, to discuss four papers on the corrosion of steel under phosphate coatings and protective finishes. Mr. H. T. Shirley, Chairman of the Corrosion Committee of B.I.S.R.A., was in the Chair. A report of the discussion at the meeting is published on pp. 177–183 of the *Journal* for February, 1953 (vol. 173).

IRON AND STEEL ENGINEERS GROUP

The Engineering Committee, which organizes the activities of the Iron and Steel Engineers Group, was composed of the following members (from May, 1952):

Chairman

Mr. C. H. T. WILLIAMS

Members

Mr. J. BAKER	Mr. E. PINCHES
Mr. W. W. FRANKLIN	Mr. W. J. POOL
Mr. J. L. GASKELL	Mr. I. S. SCOTT-MAXWELL
Mr. T. A. HOGARTH	Mr. A. TAYLOR
Mr. W. A. JOHNSON	Mr. W. UDALL
Mr. W. M. LARKE	Dr. W. S. WALKER
Mr. P. A. LEE	Mr. R. L. WILLOTT
Mr. G. S. MARTIN	Mr. C. H. WILLIAMS
Mr. J. E. OWSTON	

Ex-Officio

Capt. H. LEIGHTON DAVIES, C.B.E. (President)
Mr. JAMES MITCHELL, C.B.E. (Hon. Treasurer)
Mr. H. H. MARDON (British Iron and Steel Research Association)

Four full-day meetings, with buffet luncheons, were held during the year, three of which took place at the offices of the Institute and one at the Institution of Electrical Engineers. Reports of the discussions at the two meetings held at the Institute on Thursday 17th April and Thursday 30th October, 1952, are published in the *Journal* (Sept., 1952, and April, 1953, respectively). The report of the discussion at the meeting held on Thursday, 11th December, 1952, will be published in May. The meeting held at the Institution of Electrical Engineers was held in conjunction with the Royal Society for the Prevention of Accidents, and took place on Tuesday 4th March, 1952. It was devoted to the discussion of two papers on "Accident Prevention in Iron and Steel Works"; the two papers and the report of the discussion were published on pp. 290–308 of the *Journal* for July, 1952 (vol. 171).

A Meeting of Junior Engineers was held at Ashorne Hill, Leamington Spa, on 19th, 20th, and 21st May, 1952. The following subjects were discussed:

" Rehabilitation of a Slab Mill."
" Experiences as a Lubrication Engineer in Iron and Steel Works."
" The Use of Lubricants in the Iron and Steel Industry."
" Locomotives in Iron and Steelworks."

Three technical films were also shown, by arrangement with Shell-Mex and B.P. Ltd. A report of the discussion was circulated to all those who attended the meeting.

PUBLICATIONS

The *Journal* was, as usual, published in twelve Monthly parts, comprising volumes 170 to 172; indexes and binding cases are being prepared. Arrangements for the preparation of a comprehensive index of the *Journal* from 1932 to 1950 inclusive, in three volumes, were put in hand during 1952; the first volume, covering the period 1946 to 1950 inclusive, should be completed during 1953.

Special Reports

Bound volumes of Special Reports No. 43 (Symposium on High-Temperature Steels and Alloys for Gas Turbines) and No. 45 (Symposium on the Corrosion of Buried Metals) were issued during 1952.

The following new Special Reports were published during the year:

" The All-Basic Open-Hearth Furnace " (*Special Report No.* 46)
" Spectrographic Analysis of Low-Alloy Steels " (*Special Report No.* 47).

Bibliographies—The following Bibliography was published during the year:

" Bibliography on Wire: Manufacture, Treatment, and Properties (including References to Cold Drawing Bars) " (*Bibliography* No. 13A).

This bibliography is a continuation of Bibliography No. 13 and covers the period from 1947 to approximately the end of 1951.

During the year, Bibliography No. 13 was reprinted.

Translation Service—Translations of twenty-eight foreign technical papers (Nos. 430–457) were issued during 1952. The charge was £1 each (10*s.* for each additional copy of the same translation). The Council again wishes to thank those organizations and individuals who have kindly made translations available for inclusion in the Series. Several translations were also made for various Members at their own expense.

JOINT LIBRARY AND INFORMATION DEPARTMENT

Members, Government Departments, Research Laboratories, Universities, and other teaching establishments have continued to make great use of the facilities offered by the Library and Information Department. During the year under review, 13,448 publications were borrowed; in 1951 the figure was 11,617. The number of textbooks acquired was 475. The Council again offers thanks on behalf of the Members to the many donors who presented copies of their works to the Library.

Use of the Lending Library is a valuable privilege of membership. Books and periodicals can be borrowed, both from the Joint Library and also on application through the Librarian, from the Science Library and the National Central Library. Under certain conditions, photostat copies of documents can be obtained for Members, both at home and abroad; in 1952, 130 photostat copies and 27 microfilms were supplied to Members.

The Information Department, an important part of the service, is prepared to answer scientific and technical enquiries from members, but it is not its function to give the type of advice which comes within the field of the metallurgical consultant.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

The Institute has continued to collaborate as closely with the British Iron and Steel Research Association as in previous years. The Library, Information, and Abstracting Services are available to the Association staff and to member-firms of the Association without charge. The *Journal* continued to be the principal medium for the publication of the research results and reports of the Association. Reports of the activities of the Association appear in the News Section of the *Journal*.

The Council would again like to express their pleasure at the close collaboration existing between the two organizations.

RELATIONS WITH OTHER SOCIETIES

The excellent relations between the Institute, the Institute of Metals, and the Institution of Metallurgists were maintained during the year, as well as with other scientific and technical Societies at home and abroad. The Sheffield Metallurgical Association and the West of Scotland Iron and Steel Institute continued to collaborate closely with the Institute, as in previous years.

Affiliated Local Societies

As from 1st January, 1952, the North Wales Metallurgical Society became affiliated to the Institute. There are now eleven Affiliated Local Societies, as follows:

Cleveland Institution of Engineers
Ebbw Vale Metallurgical Society
Leeds Metallurgical Society
Lincolnshire Iron and Steel Institute
Liverpool Metallurgical Society
Manchester Metallurgical Society
Newport and District Metallurgical Society
North Wales Metallurgical Society
Sheffield Society of Engineers and Metallurgists
Staffordshire Iron and Steel Institute
Swansea and District Metallurgical Society.

Under the scheme of affiliation, the Institute and the Local Societies can be of mutual assistance without in any way diminishing their complete freedom of action. The Local Societies admit Members of the Institute to their Meetings. The Institute pays to each Local Society a capitation grant of 5*s.* per annum for each Joint Member, and any of these who are Associate Members of the Institute pay a reduced annual subscription of £1 11*s.* 0*d.*

Joint Meetings

Joint Meetings were held during the year between the Institute (including the Engineers Group) and the following Affiliated Local Societies: the Cleveland Institution of Engineers, the Ebbw Vale Metallurgical Society, the Manchester Metallurgical Society, the Sheffield Society of Engineers and Metallurgists, and the Swansea and District Metallurgical Society. Joint Meetings were held also with the West of Scotland Iron and Steel Institute and with the Sheffield Metallurgical Association.

Metallurgical Societies Overseas

The number of " Kindred " Societies overseas is now eighteen, as follows:

American Institute of Mining and Metallurgical Engineers (Associate Member age limit, 33)
American Iron and Steel Institute

American Society for Metals
Association des Ingénieurs sorti de l'Ecole de Liége (Belgium)
Associazione Italiana di Metallurgia (Italy)
Association Luxembourgeoise des Ingénieurs et Industriels (Luxembourg)
Australian Institute of Metals
Canadian Institute of Mining and Metallurgy
Indian Institute of Metals
Instituto del Hierro y del Acero (Spain)
Jernkontoret (Sweden)
Koninklyk Instituut van Ingenieurs (Holland)
Norsk Metallurgisk Selskap (Norway)
Société Française de Métallurgie (France)
Société Royale Belge des Ingénieurs et des Industriels (Belgium)
Verein deutscher Eisenhüttenleute (Germany)
Verein Eisenhütte Oesterreich (Austria)
Verein Schweizerischer Maschinen - Industrieller (Switzerland).

Members of the Institute who are Members of, or are on the staffs of members' firms of, any of the above Societies, if resident in the countries concerned, pay the following reduced subscriptions to The Iron and Steel Institute: Ordinary Members, £3 13*s.* 6*d.*; Associate Members, £1 11*s.* 0*d.*

Representation at Meetings

Mr. I. F. L. Elliot represented the Institute at the Annual Meeting of the American Iron and Steel Institute in New York on 21st and 22nd May, 1952.

Mr. K. Headlam-Morley, Secretary, represented the Institute at the meeting of the Société Française de Métallurgie in Paris from 20th to 25th October, 1952, and at the Meeting of the Verein deutscher Eisenhüttenleute in Düsseldorf from 5th to 8th November, 1952.

EDUCATION

Joint Committee on Metallurgical Education

The Joint Committee published a report on "The Education and Training of Metallurgists" in September, 1952. The following is a summary of the recommendations:

"(1) The universities are at present, are likely to be, and should continue to be, the chief source of high-grade technologists. Universities should expand their output of technologists to the greatest extent they can achieve.

(2) Employers should finance the attendance of suitable employees at post-graduate courses in universities.

(3) Formally-established methods should be adopted for training graduates immediately they enter industry.

(4) Since more technologists are needed than the universities can supply, the remainder coming from colleges of technology, a few colleges of technology should be up-graded and equipped for the education of high-grade technologists.

(5) In high-grade, technical teaching institutions, instruction in technical and elementary subjects should be physically separated from that in higher technological subjects.

(6) The present standard of staffing in technical colleges and colleges of technology should be raised.

(7) There should be active collaboration in various fields between colleges of technology and employers.

(8) The whole of the present examinations in metallurgy for non-university students should be considered with a view to their co-ordination and rationalization and the establishment of the Associateship of the Institution of Metallurgists as the normal goal for almost all such students.

(9) Further provision for sandwich courses should be made.

(10) Since it appears that operatives in the metallurgical industries are insufficiently responsive to the opportunities offered for education and training, the greater use of incentives should be examined; employers should take into account any examination successes by their employees when considering promotion and the filling of vacancies in operational departments.

(11) To achieve a balance in the education of metallurgists, all teaching establishments—including the universities—and employers are recommended to instruct students and employees in the new humanities."

The Committee stated that comments on the recommendations made in the report would be welcomed and given careful attention.

The report on "The Education and Training of Metallurgists" is the third of the series issued by the Joint Committee. Those previously published are "Recommendations on Qualifications for Entrance to University Schools of Metallurgy" (1948) and "Recommendations on University Full-Time Degree Courses in Metallurgy" (1950). All have been widely circulated to educational authorities and to those in industry who are likely to be interested.

Joint Committee for National Certificates in Metallurgy

The Joint Committee approved five new schemes for National Certificates in Metallurgy during 1952; other schemes have been revised, and schemes are also being organized in districts not yet covered, with a view to suitable courses being started as soon as possible.

The numbers of courses in operation during 1952 were:

Ordinary Certificate Courses	30
Contributory Centres with first or first and second year Courses	6
Higher Certificate Courses	18

Final examinations for the Ordinary National Certificate were held at 28 Technical Colleges and for the Higher Certificate at 16 Technical Colleges.

Comparative figures of entries and successful candidates for the seven years since the commencement of the Schemes for National Certificates in Metallurgy are as follows:

Ordinary National Certificate

Year	Technical Colleges	Number of Candidates Entered	Number of Candidates Awarded Certificates
1946	4	28	23
1947	11	102	69
1948	15	123	74
1949	21	222	107
1950	20	281	142
1951	25	314	138
1952	28	334	186*

Higher National Certificate

Year	Technical Colleges	Number of Candidates Entered	Number of Candidates Awarded Certificates
1946	...	...	...
1947	2	22	18
1948	8	38	22
1949	10	74	59
1950	11	62	42
1951	16	110	54
1952	16	132	102*

* The results of 8 candidates for the Ordinary Certificate and 17 candidates for the Higher Certificate whose records were not available in 1951 have been included in 1952.

The lists of those who qualified for Certificates and for the receipt of prizes are published in the Joint Committee's annual report.

HONOURS CONFERRED ON MEMBERS

The Council offers hearty congratulations to Members who have received honours or appointments during the year. Their names have been recorded in the News of Members columns of the *Journal*.

FINANCE

(*The Accounts for* 1952 *are appended*)

(Figures for the previous year are printed in brackets for comparison.)

General Fund

The Balance Sheet at 31st December, 1952, is presented in the usual form. The total of the Accumulated Fund, Entrance Fees Reserve, Life Composition Fund and Sundry Provisions amounted to £43,560 (£40,930).

Income for the year ended 31st December, 1952, was £69,540 (£64,686). This was largely due to an increase in receipts from Advertisements and from *Journal* Sales.

Normal expenditure was £66,213 (£61,180) and total expenditure, after including non-recurring items, was £67,111 (£63,968). The increase was due mainly to the increased cost of printing the *Journal*.

The excess of income over expenditure transferred to the Balance Sheet was £2,429 (£718).

In accordance with usual practice, a separate Publications Account is submitted, in which the expenditure on salaries and overheads is not included. The account shows that the amount transferred to the Income and Expenditure Account of the General Fund was £17,105 (£14,438).

Trust Funds

Income of the Trust Funds again exceeded expenditure, with the result that the amounts standing in the accumulated funds increased: Andrew Carnegie Trust Fund, £7,137 (£6,627); Williams Prize Fund, £777 (£749); Bessemer Medal Fund, £46 (£42); Sir Robert Hadfield Fund, £20 (£15).

Investments

The market value of the investments of the General Fund at 31st December, 1952, was £31,617 (£31,673) and, therefore, was again less than the cost, £32,982, at which they are taken into the Balance Sheet. The total figures for investments of the General and Trust Funds at 31st December, 1952, were as follows: Nominal value, £86,992 (£86,992); Cost, as included in the Balance Sheet, £76,066* (£76,066); Market value, £74,851 (£75,283).

* This figure includes £474 15*s*. 0*d*. British Transport 3% Guaranteed Stock 1978/88 at the figure of £400 at which it is included in the Balance Sheet of the Bessemer Medal Fund.

THE IRON AND STEEL INSTITUTE

BALANCE SHEET, 31st DECEMBER, 1952

General Account

1951	LIABILITIES				1951	ASSETS		
£	£		£	£	£	£	£	£
		Accumulated Fund :				*Furniture and Fittings*		Not valued
26,825		Balance as at 31st December, 1951	27,293		4,331	Stock of Paper		2,134
718		*Add* Excess of Income over Expenditure for the year	2,429					
27,543			29,722		32,982	*Investments at Cost per Schedule* (Market value at 31st December, 1952—£31,617)		32,982
250		*Less* Transfer to Reserve for Ten Year Index	250					
27,293	29,472					*Sundry Debtors and Payments in Advance :*		
		Reserve :			6,726	Sundry Debtors	8,143	
5,978		Entrance Fees Reserve : Balance as at 31st December, 1951		5,978	2,013	British Iron and Steel Research Association	1,026	
					181	Current Accounts	652	
		Current Liabilities and Provisions :			806	Deposits and Payments in Advance	1,261	
9,043		Sundry Creditors (part estimated)	5,832		2,616	*Balances at Post Office Bank and Cash in Hand*	2,866	
370		Due to Life Composition Fund	371		12,342			13,948
15		Due to Sir Robert Hadfield Medal Fund	20		2	*Joint Committee on Metallurgical Education*		428
42		Due to Bessemer Medal Fund	50					
43		Ablett Prize Fund Account	143					
207		Subscriptions in Advance	188					
1,158		Payments received in advance for Publications	1,741					
29		Due to Andrew Carnegie Research Fund	197					
		Sundry Provisions as at 31st December, 1951 :						
	850	Reserve for Ten Year Index	850					
		Transfer from Accumulated Fund	250					
				1,100				
	500	Library Account	500					
	1,000	Repairs, Decorations, and Equipment	1,000					
2,350				2,600				
3,129		Bank Overdraft	2,900					
16,386	14,042							
£49,657				£49,492	£49,657			£49,492

Special Funds

1951					1951			
£	£		£	£	£	£	£	£
		Life Composition Fund :				*Life Composition Fund :*		
5,141		Balance as at 31st December, 1951	5,309		4,939	Investments at Cost per Schedule (Market value at 31st December, 1952—£5,004)	4,939	
168		*Add* Compounded during 1952	—		370	Due from General Account	370	
5,309	5,309				5,309			5,309
£54,966				£54,801	£54,966			£54,801

We have examined the above Balance Sheet of the Institute and found it to be correct.

FINSBURY CIRCUS HOUSE,
BLOMFIELD STREET,
LONDON, E.C.2
3rd March, 1953

(Signed) W. B. KEEN & Co.,
Chartered Accountants

INCOME AND EXPENDITURE ACCOUNT FOR THE YEAR ENDED 31ST DECEMBER, 1952

EXPENDITURE

1951 £	£		£	£	£
		Staff Expenses :			
20,506		Salaries		21,303	
		Less Contributions from :			
		Andrew Carnegie Research Fund	150		
		Institute of Metals	100		
		Institution of Metallurgists	60		
		Joint Committee on Metallurgical Education	300		
		Joint Committee for National Certificates in Metallurgy	500		
		Mond Nickel Fellowships Committee	250		
1,360				1,360	
19,146				19,943	
330		National Insurance		354	
2,365		Staff Superannuation Fund and Pension		2,692	
21,841		Total Staff Expenses			22,989
		Other Administrative Expenses:			
1,817		Printing and Stationery		1,596	
635		Postage and Receipt Stamps		740	
1,300		Travelling and Entertainment : General		1,662	
1,534		Advertisement Collection		1,800	
		Meetings :			
658		General and Engineers Group	1,249		
593		High Temperature Symposium	—		
30		Buried Metals Symposium	—		
				1,249	
343		Telephone and Telegrams		382	
90		Insurances		87	
147		Audit Fee—1951		188	
42		Interest on Overdraft		13	
279		Sundry Payments		265	
7,468					7,982
		Establishment Expenses :			
1,200		Rent		1,200	
551		Rates		482	
1,563		Cleaning, Heating, Lighting, and Water		1,635	
424		Repairs and Decorations		574	
150		Office Furniture and Fittings		790	
3,888					4,681
		Publications : (*See* Publications Account)			
21,152		1951 *Journal* : Printing, Paper, etc.		24,669	
3,202		Other Publications : Printing, Paper, etc.		1,776	
272		Translations : Preparation		363	
24,626					26,808
		Library and Information Department :			
760		Library : Books and Binding		1,052	
52		Book Reviewers' Fees		54	
874		Abstractors' Fees		904	
1,686					2,010
59,509		*Carried Forward*			64,470

INCOME

1951 £	£		£	£
		Subscriptions :		
15,895		Current	16,022	
528		Arrears	515	
16,423				16,537
45		*Companies' Special Subscriptions*		—
		British Iron and Steel Research Association :		
6,000		Block Grant for Services other than Publications		6,000
		Interest on Investments (Gross) :		
1,147		General Fund	1,222	
164		Life Composition Fund	186	
1,311				1,408
54		*Post Office Savings Bank Interest*		65
		Institute of Metals :		
950		Rent and Services	950	
500		Library	500	
1,450				1,450
83		*Sundry Receipts*		54
		Publications : (*See* Publications Account)		
		1952 *Journal :*		
35,805		Revenue from Advertisements, Sales, and Amounts payable by British Iron and Steel Research Association	40,860	
2,987		Other Publications—Sales and Payments by British Iron and Steel Research Association	2,690	
386		Translations—Sales	476	
39,178				44,026
		Adjustments for 1950 :		
142		Amounts overprovided		
64,686		*Carried Forward*		69,540

£	£		£	£
59,509		*Brought Forward*		64,470
		Miscellaneous :		
1,249		Historical Investigator : Honorarium and Expenses	1,291	
174		Contributions to Affiliated Local Societies	182	
		Education :		
100		Joint Committee for National Certificates in Metallurgy ..	100	
50		Joint Committee on Metallurgical Education	50	
150			150	
		Grants :		
2		Joint Committee for Materials and their Testing	2	
25		British Standards Institution	25	
16		Parliamentary and Scientific Committee	16	
5		Association of Libraries and Information Bureaux ..	5	
50		Williams Prize Fund	—	
98			48	
1,671				1,671
		Adjustments for 1951 :		
		Amount underprovided : Indexes	121	
		Bound Volumes	24	
			145	
		Less : Overprovided Binding Cases	73	
				72
3,506		Balance carried down		3,327
£64,686				£69,540
1,074		Deficit on Special Summer Meeting in Austria	—	—
—		Cost of Entertainment of Foreign Visitors	1,151	
—		*Less* Surplus on Special Meeting in Swansea	253	
				898
1,439		Repairs and Decorations (Special)	—	—
275		Rates : Liability for 1950	—	—
718		Excess of Income over Expenditure for the year		2,429
£3,506				£3,327

£		£
64,686	*Brought Forward*	69,540
3,506	Balance brought down	3,327
£64,686		£69,540

PUBLICATIONS ACCOUNT FOR THE YEAR ENDED 31ST DECEMBER, 1952

1952 *Journal*

1951		EXPENDITURE			1951		INCOME		
£	£		£	£	£	£		£	£
16,934		Printing, Paper, and Blocks : Text and Advertisements	20,668		25,890		Advertising Revenue		29,003
250		Bound Volumes (estimated)	300			4,018	Sales : *Journal*—1952 Monthly Journal and Bound Volumes	4,253	
600		Indexes (estimated)	800			896	*Journal* Back Numbers	1,478	
250		Binding Cases (estimated)	300			740	Reprints and Blocks	1,614	
2,376		Packing and Postage	1,685		5,654				7,345
742		Reprints	916		230		Binding Cases		261
21,152				24,669	4,031		Amounts payable by British Iron and Steel Research Association ..		4,251
14,653		Balance transferred to Income and Expenditure Account		16,191					
£35,805				£40,860	£35,805				£40,860

Other Publications

	EXPENDITURE				INCOME	
£		£	£			£
305	Bibliographies	315	2,987	Sales and Payments by British Iron and Steel Research Association ..		2,690
2,897	Special Reports	1,461	215	Balance transferred to Income and Expenditure Account		—
	Balance transferred to Income and Expenditure Account	914				
£3,202		£2,690	£3,202			£2,690

ANDREW CARNEGIE RESEARCH FUND

BALANCE SHEET, 31st DECEMBER, 1952

1951		LIABILITIES		
£	£		£	£
		Amount of Original Fund, $100,000	21,241	
		Add Amount Capitalized per last Balance Sheet	4,596	
25,837				25,837
		Sundry Creditors:		
	768	Grants	523	
	10	Auditors' Fees	16	
778				539
2,600		Reserve for Travelling Scholarships—Balance at 31st December, 1951		2,600
6,627		Accumulated Fund—Balance per Account annexed		7,137
£35,842				£36,113

1951		ASSETS		
£	£		£	£
34,708		Investments at Cost per Schedule		34,708
		(Market value at 31st December, 1952—£34,787)		
		Balances at Banks:		
	358	On Current Account	447	
	9	On Deposit Account	9	
	575	Post Office Savings Bank	589	
942				1,045
163		Income Tax Recoverable		163
29		Iron and Steel Institute: Current Account		197
£35,842				£36,113

ACCUMULATED FUND, 31st DECEMBER, 1952

£		£
950	Grants	650
150	Contribution to General Fund on account of Salaries	150
10	Audit Fee	5
	1951, Underprovided	
	1952,	16
		21
5	Sundry Expenses	—
6,627	Balance carried forward	7,137
£7,742		£7,958

£	£		£	£
6,433		Balance as at 31st December, 1951		6,627
		Income for Year ended 31st December, 1952:		
	1,295	Interest on Investments (Gross)	1,317	
	14	Interest on Deposits	14	
1,309				1,331
£7,742				£7,958

WILLIAMS PRIZE FUND

BALANCE SHEET, 31st DECEMBER, 1952

1951	LIABILITIES	
£		£
2,220	Capital Value representing Market Value of £3,000 3½% Conversion Loan on 21st September, 1926, when the Fund was inaugurated	2,220
749	Accumulated Fund—Balance per Account Annexed	777
£2,969		£2,997

1951		ASSETS		
£	£		£	£
2,769		Investments at Cost per Schedule		2,769
		(Market Value at 31st December, 1952—£2,828)		
		Balance at Banks:		
	37	On Current Account	63	
	161	Post Office Savings Bank	165	
198				228
2		Sundry Debtor		—
£2,969				£2,997

ACCUMULATED FUND, 31st DECEMBER, 1952

£		£
100	Awards	100
749	Balance carried forward	777
£849		£877

£	£		£	£
723		Balance as at 31st December, 1951		749
		Income for Year ended 31st December, 1952:		
	121	Interest on 3½% Conversion Loan	121	
	3	Interest on Post Office Savings Bank	4	
	2	Interest on 3% Defence Bonds	3	
126				128
£849				£877

BESSEMER MEDAL FUND

BALANCE SHEET, 31ST DECEMBER, 1952

1951	LIABILITIES	£	1951	ASSETS	£
£			£		
	Capital Value representing Market Value of £400 4% Debentures of L. & N.W.			£474 15*s.* 0*d.* British Transport 3% Guaranteed Stock 1978/88	
400	Railway	400	400	(Market Value at 31st December, 1952—£366)	400
42	Accumulated Fund—Balance per Account Annexed	50	42	Sundry Debtor—Iron and Steel Institute	50
£442		£450	£442		£450

ACCUMULATED FUND, 31ST DECEMBER, 1952

£		£	£		£
4	Medal	6	32	Balance as at 31st December, 1951	
42	Balance carried forward	50		Income for Year ended 31st December, 1952 :	42
			14	Interest on British Transport Stock	14
£46		£56	£46		£56

SIR ROBERT HADFIELD MEDAL FUND

BALANCE SHEET, 31ST DECEMBER, 1952

1951	LIABILITIES	£	1951	ASSETS	£
£			£		
	Capital Value representing Market Value of £267 11*s.* 2*d.* 3% Savings Bonds			£267 11*s.* 2*d.* 3% Savings Bonds 1955/65	
268	1955/65	268	268	(Market Value at 31st December, 1952—£249)	268
15	Accumulated Fund—Balance per Account Annexed	20	15	Sundry Debtor—Iron and Steel Institute	20
£283		£288	£283		£288

ACCUMULATED FUND, 31ST DECEMBER, 1952

£		£	£		£
3	Medals	3	10	Balance as at 31st December, 1951	15
15	Balance carried forward	20		Income for Year ended 31st December, 1952 :	
			8	Interest on 3% Savings Bonds	8
£18		£23	£18		£23

THE IRON AND STEEL INSTITUTE

SCHEDULE OF INVESTMENTS AT 31ST DECEMBER, 1952
SHOWING NOMINAL VALUES, COST AND PRESENT MARKET VALUES

Nominal Value	Nature of Security	Market Value, 31st December, 1952	Cost
	GENERAL FUNDS OF THE INSTITUTE		
£ s. d.		£	£
2,197 7 0	3½% War Stock	1,714	2,161
1,324 7 4	3½% Conversion Loan	1,046	1,254
2,954 1 0	4% Consolidated Stock	2,570	3,176
12,242 19 3	4% Funding Loan, 1960/90	12,059	9,512
1,000 0 0	3% Defence Bonds (P.O. Register)	1,000	1,000
2,000 0 0	3% Savings Bonds, 1955/65	1,860	2,000
5,000 0 0	3% Do. 1955/65	4,650	4,972
2,000 0 0	3% Do. 1960/70	1,720	2,000
1,500 0 0	3% Do. 1965/75	1,260	1,500
4,854 2 0	British Transport 3% Guaranteed Stock, 1978/88	3,738	5,407
		£31,617	£32,982
	LIFE COMPOSITION FUND		
587 13 10	3½% Conversion Loan	464	513
205 15 6	3½% War Stock	160	212
681 13 0	4% Funding Loan, 1960/90	672	600
196 10 1	4% Do. 1960/90	194	231
2,615 11 2	British Transport 3% Guaranteed Stock, 1978/88	2,014	1,893
1,500 0 0	3% Defence Bonds (P.O. Register)	1,500	1,490
		£5,004	£4,939
	ANDREW CARNEGIE RESEARCH FUND		
	(Trustees: Sir Wm. Larke, K.B.E., The Hon. R. G. Lyttelton, and N. H. Rollason)		
6,897 14 11	3½% War Stock	5,380	6,896
800 0 0	3½% Do.	624	794
2,693 12 0	3½% Conversion Loan	2,128	2,007
3,922 14 10	3% Serial Funding Stock, 1955, (formerly 2½% National War Bonds, 1951/53)	3,933	3,998
1,793 15 0	British Electricity 3% Guaranteed Stock, 1968/73	1,507	1,762
2,166 18 7	4% Funding Stock, 1960/90	2,134	2,643
500 0 0	2½% Defence Bonds (P.O. Register)	500	500
1,250 0 0	3% Do.	1,250	1,250
750 0 0	3% Do.	750	745
17,812 10 3	British Transport 3% Guaranteed Stock, 1978/88	13,716	10,863
1,000 0 0	3% Savings Bonds, 1955/65	930	1,000
2,250 0 0	3% Do. 1960/70	1,935	2,250
		£34,787	£34,708
	WILLIAMS PRIZE FUND		
	(Trustees: Sir Wm. Larke, K.B.E., The Hon. R. G. Lyttelton, and N. H. Rollason)		
3,452 15 7	3½% Conversion Loan	2,728	2,670
100 0 0	3% Defence Bonds (P.O. Register)	100	99
		£2,828	£2,769
	SIR ROBERT HADFIELD MEDAL FUND		
267 11 2	3% Savings Bonds, 1955/65	£249	£268
	BESSEMER MEDAL FUND		
	(Trustees: Desmond Lysaght, The Hon. R. G. Lyttelton, and N. H. Rollason)		
474 15 0	British Transport 3% Guaranteed Stock, 1978/88	£366	—

(*Signed*) W. BARR,
Hon. Treasurer

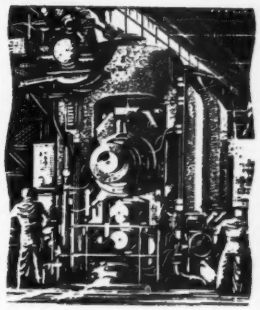
(*Signed*) K. HEADLAM-MORLEY,
Secretary

We have examined the foregoing Balance Sheets and Income and Expenditure Accounts with the Books and Vouchers of the Institute and certify them to be correct. We have also verified the balances at the Bankers and the Securities for the Investments shown above.

FINSBURY CIRCUS HOUSE,
BLOMFIELD STREET, E.C.2
3rd March, 1953

(*Signed*) W. B. KEEN & CO.,
Chartered Accountants

Control of Strip Thickness in Cold Rolling by Varying the Applied Tensions



By R. B. Sims, B.Sc., A.M.I.Mech.E., A.Inst.P., J. A. Place, B.Sc., A.Inst.P., and P. R. A. Briggs, B.Sc., A.Inst.P.

SYNOPSIS

When a rolling mill is run at constant speed, errors in gauge arise from changes in hardness, width, and thickness of the incoming material and from variations in the rolling friction. The gauge of strip may also vary when the mill speed is changed.

An automatic regulator has been built to demonstrate on a pilot-plant scale a method of controlling the gauge of material cold-rolled in a strip mill. A measure of the departure of rolled-strip thickness from the desired value is obtained from the mill load and a correction is made by varying the applied strip tensions. The controller is described in the paper, together with a number of subsidiary circuits which make the system of control suitable for production mills.

Experiments have been carried out which show that the gauge of strip may be held to very close tolerances by this method. 751

Introduction

FIGURE 1 shows the thickness variation along the length of a coil of cold-rolled steel strip. The B.S. Specification for the gauge of this class of material is to within $\pm 0{\cdot}0015$ in. of the nominal thickness. Although these tolerances are wide, almost half the length of the coil is close to or just outside these limits, and this is not considered to be exceptional. Mill operators experience great difficulty in controlling gauge closely at high strip speeds, for it is well known that if the screws are adjusted quickly a series of under- and over-corrections occur which may give rise to gauge errors greater than those that were to be corrected. The thickness profile of this coil illustrates one of the major problems of the cold-rolling industry; the speed at which it is economical to run the mill is too great for the operator to correct the variations occurring in the rolled-strip thickness due to imperfections both in the ingoing strip and in the cold-reduction process, but at the same time new markets for strip are demanding closer tolerances in gauge. The need for automatic gauge control on rolling mills is, therefore, both obvious and urgent.

The origins of the gauge variations in rolled strip have been discussed quantitatively by Hessenberg and Sims,[1] and from the relationship that they derived between mill load and rolled-strip thickness two new methods were suggested for the control of gauge, in which the gauge error signal is derived from the mill load. In one of these methods, suitable only for cold-strip mills, the correction is obtained by varying the tensions applied to the strip, and this paper describes the construction and performance of a prototype automatic regulator which was built to study in detail the specification of a design for production mills.

Notation

The following symbols are used:

F = force tending to separate the rolls, tons
F_0 = value of F when no tension is applied to the strip, tons
F_s = value of F for a pass with automatic control, tons
H = entry thickness of strip, in.
h = exit thickness of strip, in.
k = yield stress of the material, tons/sq. in.
$\bar{k}$ = mean value of k, tons/sq. in.
M = elastic constant of the mill, tons/in.
$r = 1 - \frac{h}{H}$, fractional reduction in the pass
S = minimum distance between rolls, in.
S_0 = value of S when mill is unloaded, in.
T_f = front tension force acting over section of strip, tons per inch width
T_s = initial value of T_f for a pass with automatic control
t_f = front tension stress applied to strip, tons/sq. in.

THEORY OF GAUGE CONTROL BY TENSION

In a strip mill, the rolls are forced against the wide relatively thin material by loading them at either end, usually by screws. Both the mill frame and the screws deform elastically under the reaction of the roll load, and the rolls are compressed between the loads applied by the screw and the equal and opposite roll separating force exerted by the plastically deforming material. The mill may be regarded, therefore, as a set of stiff

Paper MW/A/40/52 of the Rolling Committee of the Mechanical Working Division of the British Iron and Steel Research Association, first received 28th October, 1952, and in its final form on 27th January, 1953. The views expressed are the authors' and are not necessarily endorsed by the Committee as a body.

Mr. Sims is now Research Manager of the Davy and United Engineering Co., Ltd. Mr. Place is head of the Rolling Section and Mr. Briggs is head of the Instruments Laboratory of the Association.

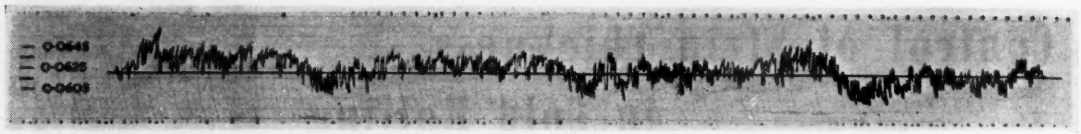


Fig. 1—Thickness profile of a coil of cold-rolled steel strip, 3·5 in. wide

springs deformed by the forces in the roll gap, and from Hooke's law it follows that the elastic distortion at any point in the mill will be proportional to the applied load. Hence, if the minimum distance between the working surfaces of the rolls is denoted by S_0 before the mill is loaded, those surfaces will be displaced relative to each other during rolling. If, at a load F, the surfaces are separated by a distance S, then by simple elastic theory:

$$F = M(S - S_0),$$

where M is a factor of proportionality termed the elastic constant of the mill.

The material will leave the rolls at a thickness equal to the distance separating the working surfaces of the rolls; therefore, if h denotes the thickness of the rolled strip and the elastic recovery of the material is neglected,* $S = h$ and

$$F = M(h - S_0) \quad \ldots\ldots\ldots\ldots\ldots\ldots (1)$$

If the roll load and roll setting S_0 do not vary during the pass, it follows from equation (1) that the strip will be rolled to a constant thickness.

When the roll setting is not altered, the load may be held constant only by adjusting the magnitude of one of the rolling variables by an amount sufficient to counteract any tendency for the load to vary with changes in the width, thickness, and hardness of the incoming strip, and in strip friction and mill speed. The only variable which may be so controlled is the tension applied to the strip, and it has been shown[2] that, as tensions are increased, the rolling load is decreased.

In the automatic gauge-controller based on these principles, the roll load corresponding to the desired strip thickness is first obtained. From equation (1), differences in the load from this value are proportional to the variation in gauge of the rolled strip, and form the error signal of the controller. As the roll load increases, the resultant error signal is arranged to increase the strip tension, which in turn reduces the roll load until the strip returns to gauge. The control loop, which begins at the loadmeter and supplies the amplified error signal to the reel motor, is completed by the strip and mill frame back to the loadmeter.

EXPERIMENTAL MILL AND CONTROLLER

The 10 in. × 10 in. mill at Sheffield University was used for the experiments. In fitting the closed loop controller it was decided to alter the mill and its drive and instrumentation as little as possible. The circuits described below were constructed within this limitation, and although they do not necessarily represent the best method of applying gauge control to a mill, they illustrate the principles involved and the performance that might be expected in an industrial design.

The mill is an old 2-high design. The non-reversing electrical drive was modernized in 1944, when the coiler gear was fitted together with Brown-Boveri regulators to maintain constant strip tension. Details of the drive have been given by Ford[3] in the original description of the mill, and only a brief description will be given here of the parts that were modified to take the controller.

The mill speed is varied by the Ward–Leonard system of control. A single main generator is driven from a slipring induction motor, and the mill speed is varied up to 150 ft./min. by varying the generator e.m.f. and from 150 ft./min. to 300 ft./min. by field weakening of the mill motor. The maximum speed is obtained only when the mill is lightly loaded. The wind-reel motor is supplied with a potential difference proportional to the mill speed by connecting it across the main generator in series with a supplementary booster motor-generator excited from a small generator or pilot exciter in the main drive. The original circuit is shown diagrammatically in Fig. 2, and the circuit after modification to take the closed loop control is shown in Fig. 3. The booster field energized by the pilot exciter is unchanged; the running tension is obtained from the field which formerly provided the IR compensation to the reel system, and the field

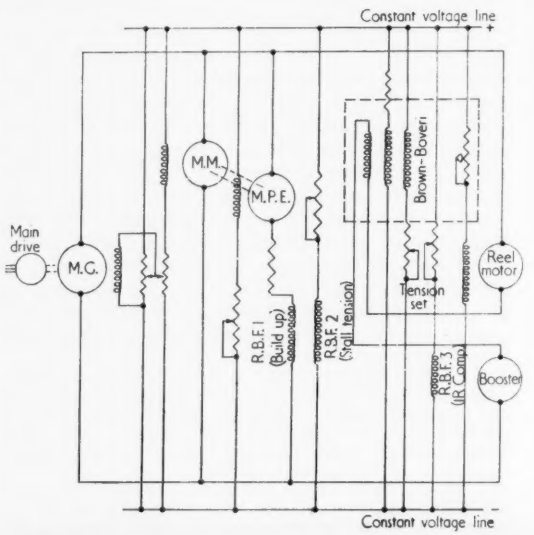


Fig. 2—Circuit of mill drive before modification

* If the rolls are separated by a distance $S = h$ and, due to elastic recovery, the strip emerges at a thickness $h(1 + \epsilon)$, the strip will be rolled to a thickness $(h + \Delta h) \times (1 + \epsilon')$ when the roll separation increases to $S + \Delta S$. When a strip is rolled with gauge control, Δh is small and $(1 + \epsilon')$ is very nearly equal to $(1 + \epsilon)$, so that during the entire pass the elastic recovery of the strip may be allowed for by a factor $1 + \epsilon$. As ϵ rarely exceeds 0·0002, it will be neglected in the general discussion, to preserve simplicity.

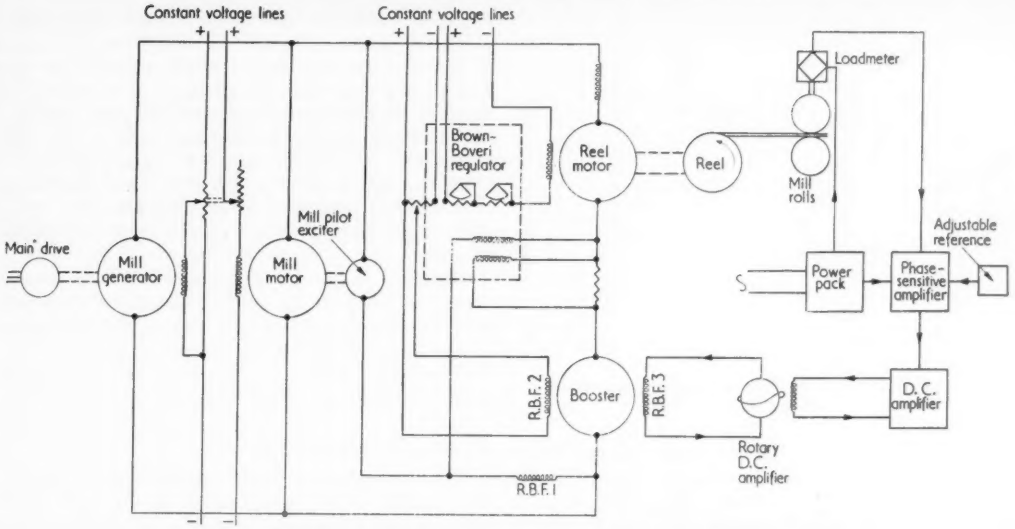


Fig. 3—Main circuit, after modification to take closed loop control

which gave stall tension in the original scheme now carries the amplified error signal of the controller.

Since the front tension applied to the strip is completely controlled by the mill loading in this form of automatic gauge control, there is no need, in principle, to correct the output of the reel motor for changes in coil diameter. An approximate correction has been incorporated,* however, to maintain the same range of control over the errors in gauge along the entire length of the coil. This is achieved by making the Brown-Boveri regulator sensitive to changes in the back e.m.f. of the reel motor. The pilot exciter attached to the main mill motor is used as a reference voltage, and the regulator controls the reel-motor field to maintain the back e.m.f. of the reel motor at a constant proportion of the pilot-exciter e.m.f., which varies directly with mill speed. This build-up control is thus nearly independent of the speed of rolling.

The mill load was measured by two B.I.S.R.A.-type loadmeters[4] placed between the mill screws and the top of the roll-neck bearing chocks. The load-sensitive element in this type of loadmeter is a short steel column on which is bonded a Wheatstone network of electric resistance strain gauges. When load is applied, the column is strained and the network is unbalanced. The potential difference across the output terminals of the network is directly proportional to the mill load. Two such networks were bonded to each of the loadmeters used in these experiments; one set was supplied with stabilized D.C. at a potential difference of 100 V. and was used to indicate the load on each screw, the other provided a signal to the controller and will be termed the 'signal network.' The total load on the mill was obtained by connecting the input and output of each of the signal networks in parallel.[4] This composite network was supplied with A.C. at 500 cycles/sec., and the output terminals connected to the electronic amplifier of the main controller.

* This adaptation of a Brown-Boveri controller to correct for changes in coil diameter is due to Dr. L. N. Bramley, whose assistance was invaluable in designing and installing the main controller.

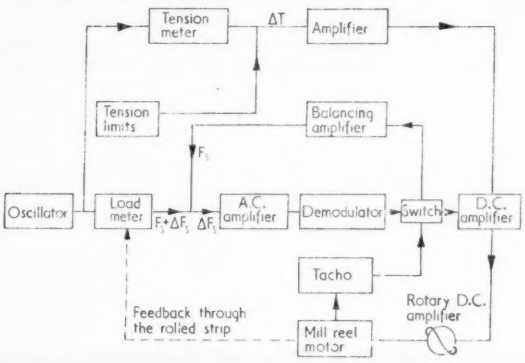


Fig. 4—Block diagram of controller circuit

Main Controller

In a pass with automatic control, the coil is held in the decoiler and threaded through the rolls on to the wind reel. The strip is then given an initial tension T_S at threading speed and the rolls are brought together until the required thickness is obtained, using a contact micrometer to gauge the strip continuously. The mill load at this point, which must be held constant by the main controller, will be denoted by F_S (*see* Fig. 4). The output from the signal networks into the electronic amplifier will be proportional to F_S, and to derive the error signal an equal and opposite reference signal is supplied to the amplifier, so that its output is reduced to zero. The control loop is closed by connecting the amplifiers to the reel-motor circuit, and the mill is accelerated to its normal running speed. The mill screws cannot be adjusted after the loop is closed without changing the rolled-strip thickness.

From equation (1) it can be seen that an error in the gauge of the rolled strip will be associated with

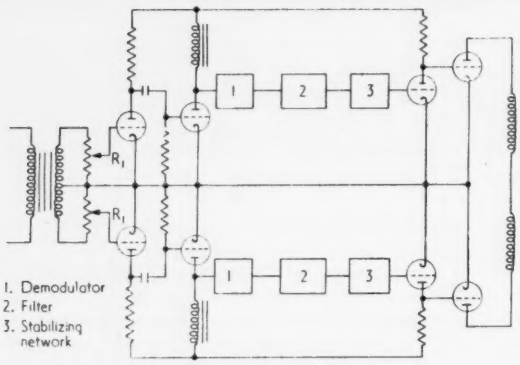


Fig. 5—Circuit of amplifier of main controller (schematic)

a change in the mill load. If the gauge changes from h to $h + \Delta h$, the load will increase from F_S to $F_S + \Delta F_S$, where $\Delta F_S = M\Delta h$. The input to the amplifier from the signal gauges will also increase, but since the reference signal remains unchanged the resultant input to the amplifier will be proportional to ΔF_S. This change in load may be either negative or positive, so that the A.C. amplifier must be phase-sensitive. The amplified error signal is then rectified and fed into a D.C. amplifier into which are connected various auxiliary controls necessary to make the system practicable for production mills. The main power amplification of the signal is obtained by a rotary D.C. amplifier, which energizes one of the booster fields. An increase in load therefore increases the e.m.f. developed across the armature of the booster and produces an increment of current in the reel motor, and, as the reel speed cannot increase, this increment increases the torque output of the motor and hence the tension in the strip.

The mill load $F_S + \Delta F_S$ will decrease as the tension is increased, so that the error signal generated by the loadmeters will be diminished. It must be emphasized, however, that the error signal cannot be reduced to zero, as the increased strip tension, which is reducing the load, is itself proportional to the error and would disappear if the error signal from the loadmeters vanished. The controller, therefore, is a closed loop system of zero order. If the rolled-strip thickness changes suddenly, the initial error signal ΔF_S is eventually reduced by the controller to $\Delta F_S/K$, where K is the static gain factor of the closed-loop system and in principle may be as large as required. This small residual error is then used to excite the control. As an example, the thickness of the coil in Fig. 1 varies by $\pm 0{\cdot}002$ in. about the mean value; if the thickness is to be held to $\pm 0{\cdot}0002$ in. along its length, a controller with a gain of 10 would be required. The factors affecting the choice of K are discussed below.

Figure 5 is the circuit diagram of the amplifier, which is followed by a phase-sensitive demodulator and a push-pull D.C. output stage. The amplification factor may be varied by the potentiometer R_1, and the control windings of the rotary amplifier form the anode load of the output stage.

Stability of the System

The evaluation of the criterion of stability for the main controller is complicated by the fact that the error signal is both generated and corrected in the roll gap, and the roll-gap transfer function is not simple. When the error signal varies sinusoidally, the loop transfer function of this controller may be written in the form

$$KG(D) = \frac{K}{[1+T_1(D)][1+T_2(D)+T_3(D^2)][1+T_4(D)][1+T_5(D)]} \quad \ldots(4)$$

where D is the differential operator $\partial/\partial t$ and the time constants T_1–T_5 refer to the following lumped circuits:

- T_1 = time constant of the circuit comprising the armature of the rotary amplifier and the booster field
- T_2, T_3 = complex time constants involving the electro-mechanical system of the armatures of the main generator, booster, and reel motor, together with the inertias of the reel motor, coil, and coiler drum and the virtual damping exerted by the roll gap due to the plastically deforming strip
- T_4 = time constant of the output stage of the electronic D.C. amplifier and the control windings of the rotary D.C. amplifier
- T_5 = quadrature time constant of the rotary D.C. amplifier.

In the closed loop fitted to the experimental mill, $T_2 \gg T_3$ and it was possible to achieve a satisfactory degree of stabilization by means of the passive integro-differential stabilizing network shown in Fig. 6. Other controllers that have been designed have not shown this property, and stabilization has been obtained by

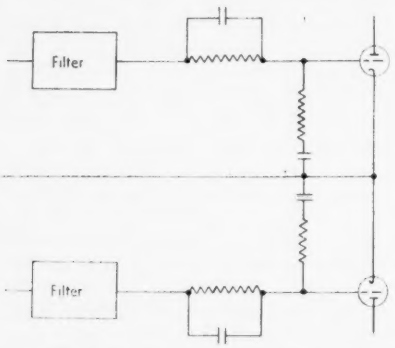


Fig. 6—Integro-differential stabilizing network for main controller

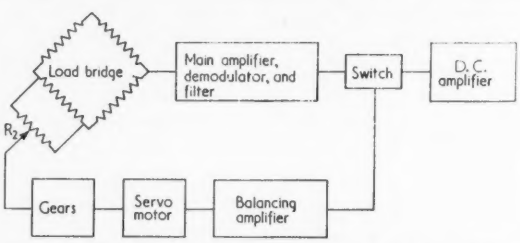


Fig. 7—Auxiliary balancing servomechanism in main controller

feedback between the elements of the loop, using a passive RC coupling.

The static gain K and the time constants T_2 and T_3 are functions of the rolling conditions, particularly the rolled-strip thickness, so that a simple estimate of the degree of gauge correction possible cannot be given without specifying the conditions in each pass. The addition of the integro-differential network resulted in a considerable increase in the static gain and an improvement in the stability and response of the main controller. The value of K was chosen so as to give an overshoot factor of 1·3 in response to unit step function when rolling the thinnest material.

Setting of the Reference Signal

On p. 344 the error signal is derived from a comparison between the load signal and a reference signal. In practice, however, it is more convenient to balance the Wheatstone network of signal gauges on the load-meter at the load F_S. A further increase in load will increase the bridge output, whereas a decrease in load will produce an output in the opposite phase. It is not easy to balance the signal gauge network by hand, as the amplification of the error signal is high; for instance, the gain of the electronic amplifier in the present prototype controller is of the order of 10^5. An auxiliary servomechanism, shown in Fig. 7, has therefore been incorporated in the main controller to perform this task for the mill operator. When the resetter control is closed, the electronic amplifier of the main controller is isolated from the rotary D.C. amplifier and the output from the demodulator is connected to an amplifier, which in turn excites a small split-field D.C. servomotor coupled through gearing to the potentiometer R_2. The motor will rotate until the output of the amplifier is very nearly zero. A passive differential network is used to stabilize this auxiliary servomechanism.

Tension-Limiting Controls

The error signal proportional to ΔF_S which is used to excite the booster increases or decreases the initial tension T_S applied to the strip before the control loop is closed. If ΔF_S is large, the tension may increase until the strip snaps, or the tension may be relaxed to zero, depending on the sign of the error. Either eventuality must be avoided in production, hence auxiliary controls have been designed which permit the tension to increase or decrease to a specified upper or lower limit, when no further change in tension occurs whatever the additional change in error signal.

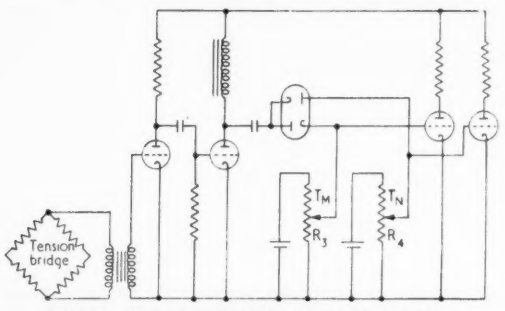


Fig. 8—Tension-limiting control circuit

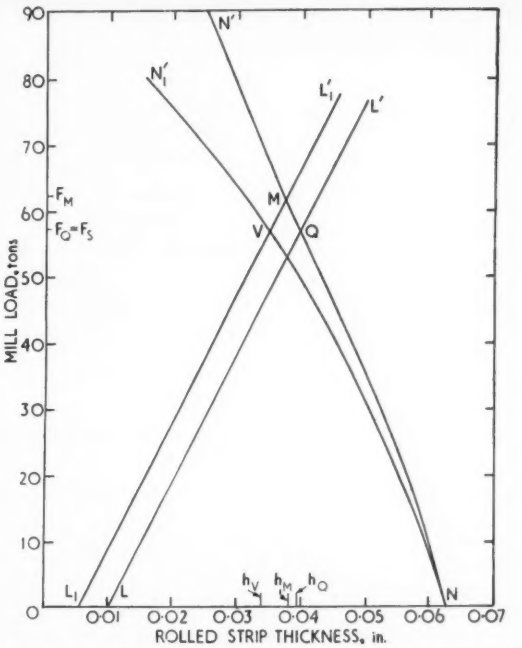


Fig. 9—An F/h diagram, showing effect on gauge of changes in thickness of oil film

In the control system adopted for the prototype, an e.m.f. proportional to the strip tension was obtained from the tensionmeter described by Rankine, Bailey, and Stanton;[5] this e.m.f. was compared with two e.m.f.s proportional to the required upper and lower tension limits, which will be denoted by T_M and T_N, respectively. When the output from the tensionmeter exceeded T_M or fell below T_N, the difference ΔT was amplified and arranged to saturate the output stage of the D.C. amplifier. The control circuit is shown in Fig. 8. The e.m.f.s proportional to T_M and T_N were set by the rheostats R_3 and R_4, which were calibrated in terms of applied strip tension.

These control circuits are secondary closed-loop controls inside the main controller, and it is essential to stabilize them separately. They include all the time lags in the main loop (*see* Fig. 4), together with the lags incurred in amplifying the e.m.f. proportional to ΔT. Moreover, when considering the stability on the Nyquist diagram the choice of the degree of stability must be influenced by the fact that, if overshoot occurs in saturating the main amplifier, these secondary controllers may become unstable because of the on-off nature of the signal.

Correction for Change in Oil-Film Thickness in Roll-Neck Bearings

Many large rolling mills are equipped with hydrodynamically lubricated bearings, and it is known that as the rotational speed of the mill is increased the oil film separating the journal from the bearing shell is also increased. This effect has been discussed by Sims,[6] who has shown that the change is equivalent to an alteration in the roll setting. The correction applied by the controller may be readily explained

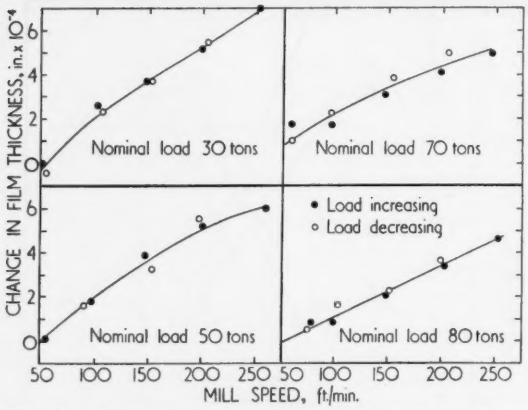

CHANGE IN FILM THICKNESS, in. x 10⁴
Nominal load 30 tons
Nominal load 70 tons
Nominal load 50 tons
Nominal load 80 tons
• Load increasing
○ Load decreasing
MILL SPEED, ft./min.

Fig. 10—Change in oil-film thickness in mill bearings with speed

when the mechanism of the effect is referred to the F/h diagram, the construction of which has been described by Hessenberg and Sims.[1] On the F/h diagram given in Fig. 9, let LL' represent the elastic equation of the mill at threading speed and NN' the plastic equation of the material. The speed effect in cold rolling due to changes in the strip friction will not be considered here,* so that as the mill is accelerated only the line of the elastic equation will shift from LL' to L_1L_1', the distance LL_1 representing the change in thickness, Δx, of the oil film. The point of intersection of the elastic and plastic lines will shift, therefore, from Q to M, and the thickness of the rolled strip will decrease from h_Q to h_M. From the geometry of the figure it will be seen that $h_Q - h_M < \Delta x$. If the pass is rolled with the control loop closed, the plastic equation will be changed to NN_1', since the controller will alter the strip tension to maintain the mill load constant. The point of intersection is now at V and the strip will be rolled to a thickness h_V, so that $h_Q - h_V = \Delta x$.

The change in the thickness of the oil film with mill speed in the experimental mill is shown in Fig. 10. This data is reproduced from the paper by Sims,[6] where details are given of the roll-neck bearings. The change in thickness is not linear and depends only on the mill speed and roll load. In tests on other mills in which the diametral bearing clearances were less, the change in film thickness with speed closely approached linearity. It has been possible to achieve a correction for this effect, therefore, by supplying the main controller with a signal proportional to mill speed.

PERFORMANCE OF THE AUTOMATIC GAUGE CONTROLLER

Control of Speed Effect

It a is characteristic of the cold-rolling process that as a mill is accelerated the rolled strip thickness is decreased, and it increases again as the mill is slowed down. The effect may be large; Stoltz and Brinks[8] found that a strip rolled to 0·013 in. on a tandem mill at a threading speed of 400 ft./min. decreased in gauge to 0·0088 in. as the mill approached its normal

* Sims and Arthur[7] have shown that the speed effect due to frictional changes may be suppressed by using colloidal graphite as a lubricant Hence this assumption, which greatly simplifies the discussion, is not unrealistic.

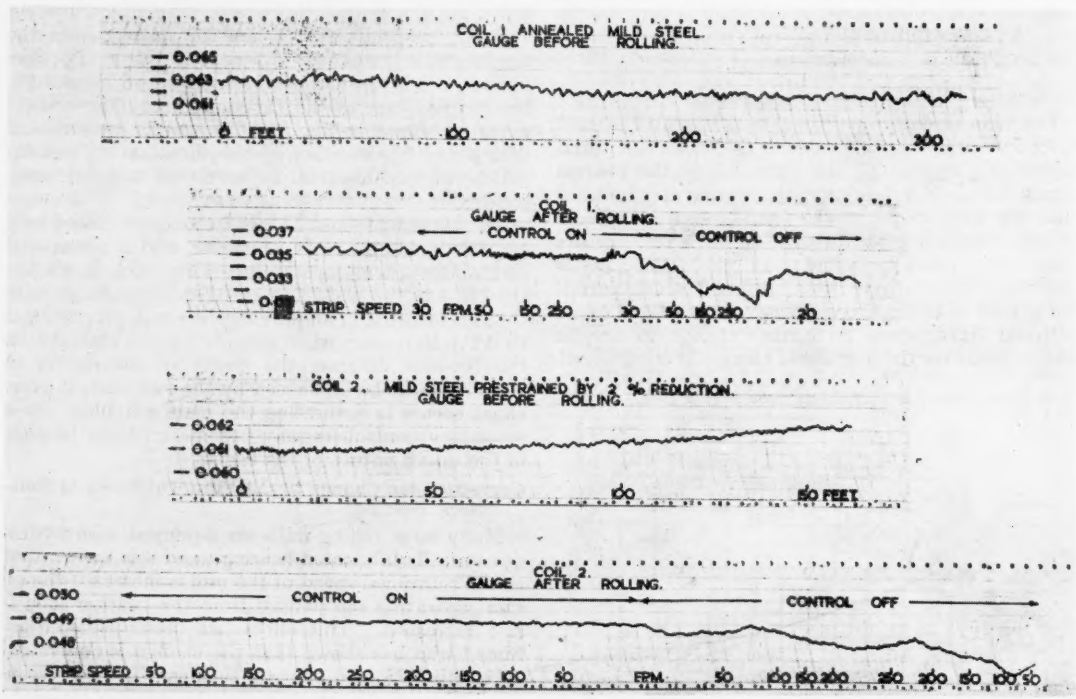


Fig. 11—Thickness profiles: (*a*) and (*b*) Coil 1; (*c*) and (*d*) Coil 2, illustrating the control of the speed effect

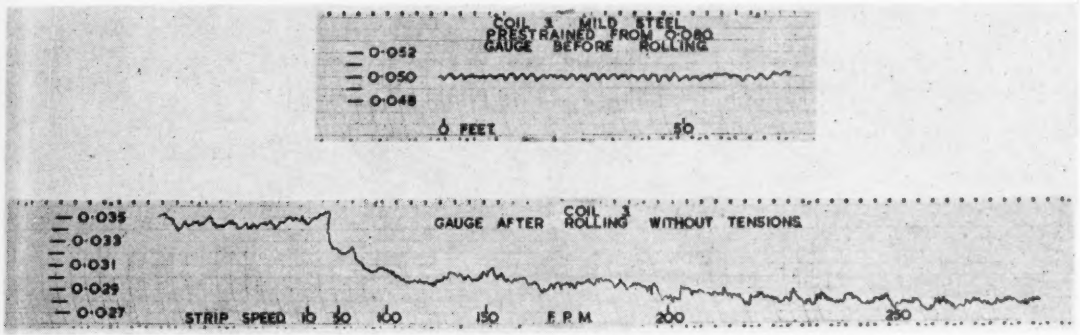


Fig. 12—Thickness profiles of Coil 3, showing true speed effect

working speed of 3500 ft./min. Two factors contribute to the effect[6, 7]: a change with speed in the coefficient of friction between the rolls and the strip, and, in mills equipped with hydrodynamically lubricated roll-neck bearings, a change in the thickness of the oil film. The off-gauge due to frictional changes in the roll gap may be reduced by the automatic controller, as it is caused solely by the changes that occur in the rolling load as the friction varies. The effect of changes in the oil film is discussed on p. 347.

Experiments were carried out with the mill rolls fitted with roller bearings and, subsequently, with hydrodynamically lubricated bearings. The thicknesses of all coils were recorded with a contact strip micrometer. In assessing the results the periodic fluctuations in gauge have been neglected; they are due to roll eccentricity, and are discussed on p. 353.

Mill Fitted with Roller Bearings—A coil of mild-steel strip was rolled at 30 ft./min. with the control loop closed. The mill was accelerated to its maximum speed and then decelerated. Without stopping the mill the control loop was disconnected and the mill accelerated again and returned to 30 ft./min.

The variation in thickness along an annealed coil (Coil 1) before rolling is shown in Fig. 11*a*, and the rolled strip thickness after the experiment is shown in Fig. 11*b*. The main controller was not fitted with the stabilizing network for this experiment, and the static gain was, therefore, reduced to keep the controller stable. Without the controller in circuit the gauge of the strip decreased by 0·004 in., but with it the gauge was held to within $\pm$ 0·0003 in.

The experiment was repeated on Coil 2, which was prestrained by rolling down from 0·063 in. to 0·061 in. The thickness profile before and after the pass is shown in Figs. 11*c* and *d*. With the controller in circuit the gauge has been held to within $\pm$ 0·0001 in. and without control it decreased by nearly 0·002 in. For this pass the controller was fitted with the stabilizing network, and the static loop gain increased, resulting in a greatly improved performance.

Two points of interest may be noted. In rolling the first coil with the controller at low gain, most of the error in the rolled strip occurred near the speed of 150 ft./min., because saturation in the booster reduced still further the overall gain of the loop. A variation in gauge at about 150 ft./min. is not perceptible in Coil 2 rolled with the controller

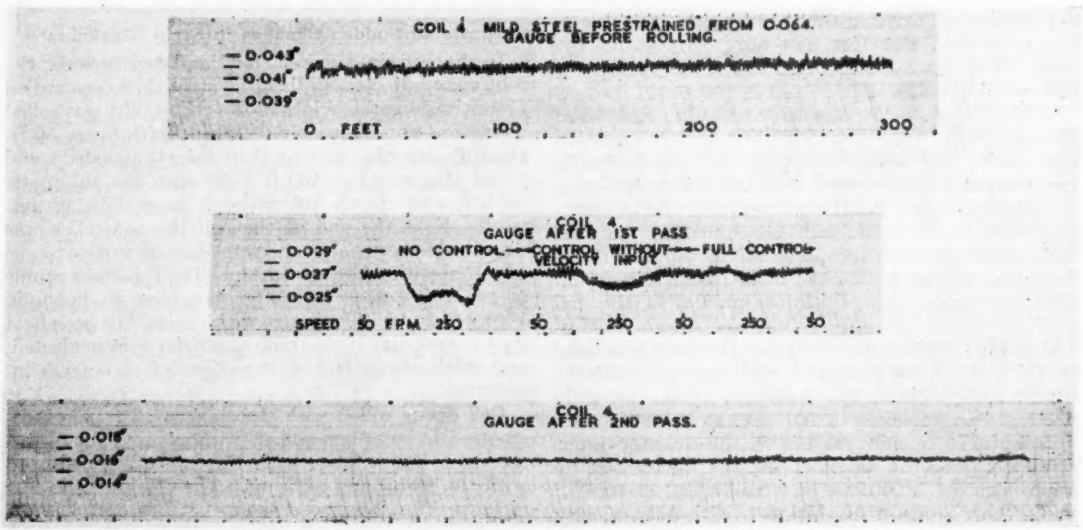


Fig. 13—Thickness profiles of Coil 4, rolled with mill fitted with hydrodynamically lubricated bearings: with and without correction for oil film thickness

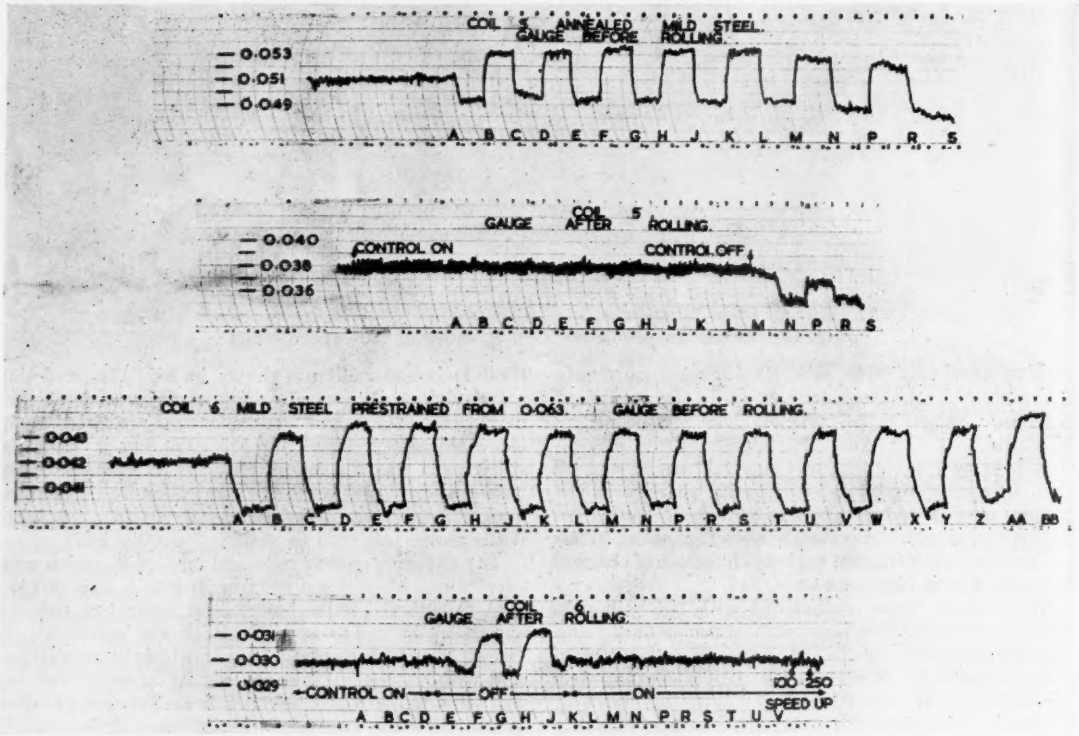

COIL 4 ANNEALED MILD STEEL
GAUGE BEFORE ROLLING.
0·053
0·051
0·049
A B C D E F G H J K L M N P R S
COIL 5
GAUGE AFTER ROLLING.
0·040
0·038
0·036
CONTROL ON
CONTROL OFF
A B C D E F G H J K L M N P R S
COIL 6 MILD STEEL PRESTRAINED FROM 0·063. GAUGE BEFORE ROLLING.
0·043
0·042
0·041
A B C D E F G H J K L M N P R S T U V W X Y Z AA BB
COIL 6
GAUGE AFTER ROLLING.
0·031
0·030
0·029
CONTROL ON
OFF
ON
100 250
SPEED UP
A B C D E F G H J K L M N P R S T U V

Fig. 14—Thickness profiles of Coils 5 and 6, showing effect of sudden changes in entry thickness of strip

at high gain. Secondly, it will be seen from Figs. 11*b* and *d* that the changes in gauge which occur when the mill is accelerated without the automatic gauge control are not repeatable. The speed effect when a coil of mild steel is rolled without tensions in the experimental mill is shown in Fig. 12. The thickness of the accelerating strip has been modified by changes in the electrical drive causing the tension to fluctuate. The controller, therefore, not only corrects for the speed effect but also compensates for the vagaries of the electrical drive.

Mill Fitted with Hydrodynamically Lubricated Bearings—The cycle of acceleration and deceleration was carried out without stopping the mill, first with the controller disconnected from the drive, then with the controller in circuit but without oil-film compensation, and finally with the controller supplied with an input signal proportional to the mill speed from the tachometer on the main motor, giving oil-film correction. The thickness profiles of the coil (Coil 4) before and after rolling are shown in Figs. 13*a* and *b*. Without the controller the strip decreased in thickness by 0·0026 in.; with the controller in circuit but without the velocity input the strip decreased in gauge by 0·0012 in. at the maximum speed of 270 ft./min., and with the velocity input connected into the amplifier the strip decreased by approximately 0·0004 in. at a mill speed of 100 ft./min. The decrease in thickness with the velocity input connected is due to the non-linearity of the changes in the oil-film thickness with speed; it is considered that the variation could be reduced by improving the bearing design.

The strip was given a second pass at 100 ft./min. to remove the irregularities in thickness due to the experiments on the speed effect. In this pass the strip thickness, shown in Fig. 13*c*, has been held to 0·016 in. $\pm$ 0·0002 in.

Response to Sudden Changes in Strip Thickness

In the first pass on coils for these experiments, the rolls were adjusted so that the strip thickness varied abruptly along the length of the coil. Coil 5 was rolled with steps approximately 0·004 in. thick every 50 ft. along its length. It was then bright-annealed* and rolled at a speed of 100 ft./min. with the automatic controller in circuit but without the stabilizing network. Near the end of the coil the controller was disconnected, to demonstrate the changes to be expected without gauge control. The thickness profile of the coil before and after the pass is shown in Figs. 14*a* and *b*. The gauge with control has been held to $\pm$ 0·0003 in. if the mill eccentricity is neglected, and without control a variation of $-$ 0·0025 in. occurred.

Coil 6 was rolled with steps every 12 ft. along its length. It was then rolled, without annealing, at a speed of 100 ft./min. with the controller equipped with the stabilizing network. The thickness profiles given in Figs. 14*c* and *d* show that not only has the

* The authors are grateful to Messrs. Arthur Lee and Co., Ltd., who carried out all the bright-annealing.

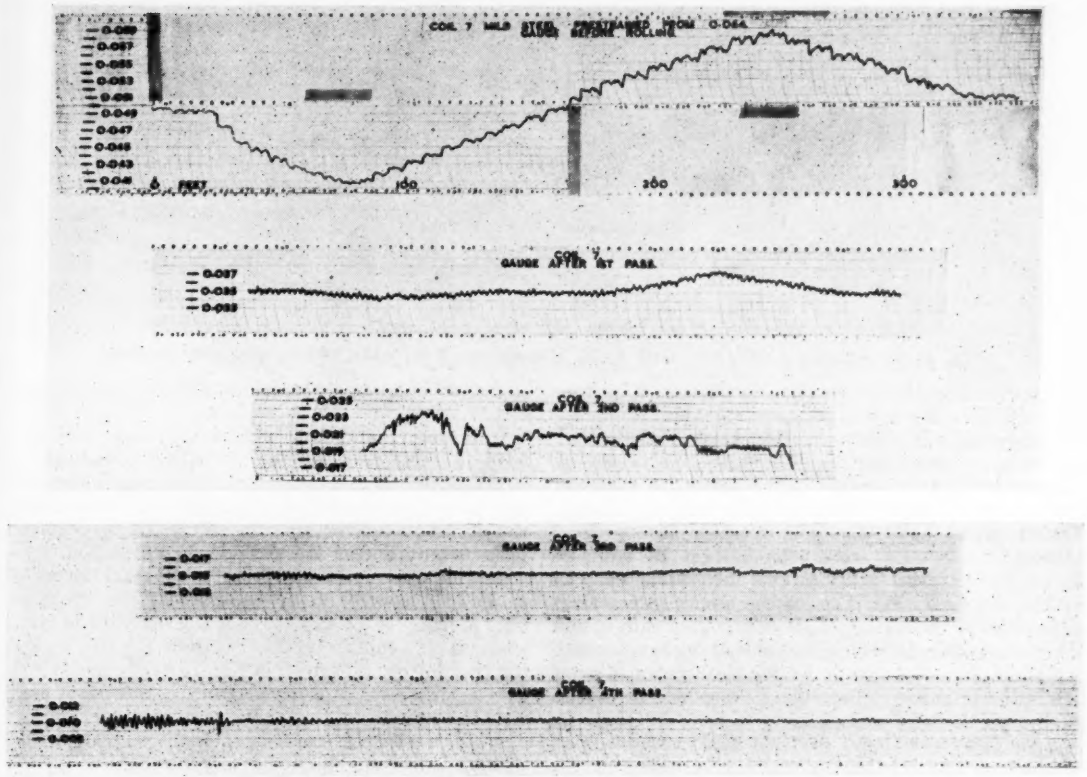


Fig. 15—Thickness profiles of Coil 7, rolled with controller fitted with tension limiting circuits

gauge been held more closely but there is no tendency to amplify the eccentricity pattern as is the case when the controller is without the network. At the point marked *V* on this coil, the mill was accelerated and the remainder of the coil was rolled at maximum speed. Both the speed effect and the changes in entry thickness have been controlled.

Performance of Limiting Tension Controllers

For this experiment a first pass was taken on Coil 7 in which the rolls were adjusted to give an approximately sinusoidal variation in gauge of $\pm 0{\cdot}009$ in. along the length of the coil. It was then rolled with automatic control with the tension limits set to $T_M = 1{\cdot}9$ tons and $T_N = 0{\cdot}6$ tons. The strip was held to gauge until the limiting controls operated and then no further correction occurred (Figs. 15*a* and *b*).

Other experiments (not connected with the present discussion) were made on this coil and after a third pass the thickness profile, shown in Fig. 15*c*, varied between 0·024 in. and 0·018 in. along its length. A pass was then taken with the automatic control loop closed and the strip rolled to 0·015 in. This provided a searching test for the limiting tension controllers, as the incoming gauge fluctuated rapidly. The profile of the strip (Fig. 15*d*) shows clearly the parts of the coil where saturation occurred.

Finally, the strip was rolled down to 0·010 in. and the gauge held to within 0·0001 in. along a length of some 1500 ft. (Fig. 15*e*). At the beginning of the pass the controller was executing a sustained oscillation and was prevented from breaking the coil only by the tension limiting system. This oscillation occurred at a static gain below the value at which instability occurs in thicker strip, and illustrates the dependence of the static gain of the loop, for a given overshoot factor, on strip thickness.

Rolling Soft Metals

The performance of the controller so far has been demonstrated with mild steel as the rolling material; it will, however, control gauge in coils of any material capable of being rolled with tension. The thickness profiles of a coil of 2S aluminium before and after rolling with automatic control are shown in Fig. 16. This material is one of the softest of those normally rolled in a strip mill, and its gauge has been held as closely as in the mild-steel coils.

MANUAL GAUGE CONTROL

The closed-loop controller described above may be rearranged to include the mill operator as one of its elements. In this case, the departure of the roll load from the pass reference ΔF_S is amplified and presented on a centre-zero microammeter. To keep the strip on gauge, the operator must vary the applied tension by hand so that the pointer of the meter remains in the central position. A controller of this type has been constructed, and the circuit is shown schematically in Fig. 17.

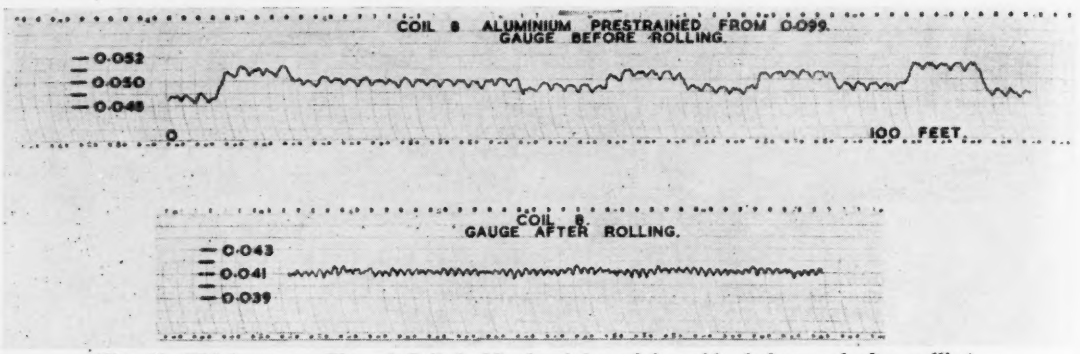


Fig. 16—Thickness profiles of Coil 8, 2S aluminium 3 in. wide, before and after rolling

The problem in servomechanics presented by this form of controller differs substantially from the automatic system. The fact that the tension is adjusted by varying a resistor implies that an element of integration is incorporated in the circuit, so that the error in gauge may, in principle, be reduced to zero. On the other hand, the mill operator as a control element introduces an involved form of transfer function,[9, 10] which complicates a theoretical analysis of the loop stability. One of the resulting practical difficulties is the setting of the static loop gain of the controller. This is determined chiefly by the amount by which the operator will alter the tension during his initial impulse on perceiving a sudden movement of the pointer of the indicating meter. The gain in the amplifier has, therefore, been made variable, and if the error signal is too large to be presented on the centre-zero meter a relay operates which prevents overloading. The signal is then transferred to a less sensitive meter, which indicates to the operator the sense and relative magnitude of the signal.

Figure 18 shows the results for a coil rolled with steps approx. 0·0005 in. thick every 14 ft. along its length. After rolling, the coil was bright-annealed and given a pass at 50 ft./min. with the manual gauge controller connected in the reel-motor circuit. A front tension force of 2·6 tons was applied initially and was decreased as the entry thickness of the strip decreased, until the strip tension was lost. The thickness of the rolled strip was held to within $\pm$ 0·0002 in. along its length until the tension became zero.

It has been found that the operator cannot maintain strip to this tolerance at strip speeds much above 100 ft./min.; in one experiment, in which the mill was accelerated to 300 ft./min. with the controller in circuit, an overcorrection of nearly 0·002 in. occurred momentarily during the period of acceleration.

An important consideration in manual forms of closed-loop control is the training of the operator. This is likely to be expensive if the operator is given training on a mill in production, and for this reason the training circuit shown in Fig. 19 has been devised. The error signal is derived from a record of load variations made on a magnetic wire recorder, and the operator varies the e.m.f. in a circuit of passive networks, which simulates the servomechanism on the mill and provides a contra-signal to the error.

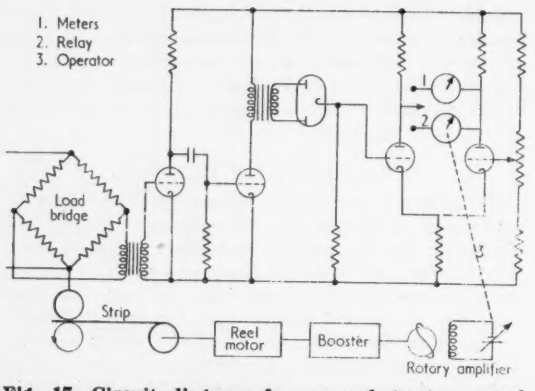


Fig. 17—Circuit diagram for manual gauge control (schematic)

RANGE OF STRIP TENSIONS REQUIRED FOR CONTROL

Calculation of Range of Tension

The reel motors of a rolling mill are costly items of capital equipment, and in drawing up the specification of a mill to be fitted with the tension method of gauge control the power of the reel motor must be calculated with care, to avoid unnecessary expense. Moreover, in applying the control to existing mills it should be verified that the magnitude of the errors that may be corrected by the tension available is greater than that of the largest error likely to occur in the run-of-the-mill material. An equation can be derived from which an estimate may be made from

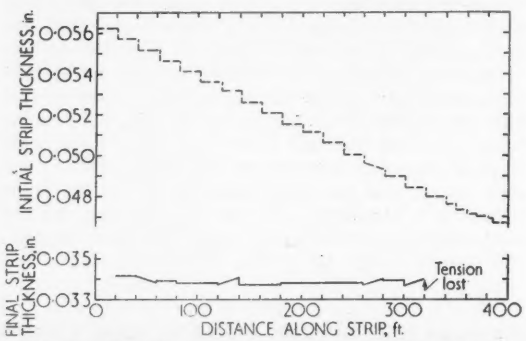


Fig. 18—Experiment with manual control of gauge

the known variations in gauge in the strip before rolling. Errors may occur, however, from changes in hardness, strip width, or friction, and these may produce larger variations in the rolled-strip thickness than those due to changes in entry thickness. The problem should always be investigated carefully, therefore, using the F/h diagram together with one of the accurate theories of rolling. The equations below are intended to serve only as a rapid and approximate method of calculating tensions and levels of correction.

Variation of Tension with Entry Thickness

The approximate relationship[2] between roll load and applied front tension is

$$F = F_0\left(1 - \frac{t_f}{2\bar{k}}\right) \quad \ldots\ldots(2)$$

so that

$$\frac{\delta F}{\delta H} = \frac{h}{H^2}\left(1 - \frac{t_f}{2\bar{k}}\right)\frac{\delta F_0}{\delta r}.$$

In practice, $\delta^2 F_0/\delta r^2$ is small and $\delta F_0/\delta r$ may be taken as a constant over a small range of reductions. Thus, writing $\delta F_0/\delta r = c$,

$$\frac{\delta F}{\delta H} = \frac{ch}{H^2}\left(1 - \frac{t_f}{2\bar{k}}\right), \quad \ldots\ldots(3)$$

where c is some function of the rolling variables.

It is assumed that control of gauge is effected by front tension only, although back tension could equally well be used. Let the entry strip thickness be H for a front tension t_f and load F; then if the thickness increases to $H + \Delta H$ the load will increase instantaneously to $F + \Delta F$. When the resulting change in exit thickness is controlled by an increase in forward tension Δt_f, then (from the principle of the method):

$$F = F_0\left(1 - \frac{t_f}{2\bar{k}}\right) = \left(F_0 + \Delta F_0\right)\left(1 - \frac{t_f + \Delta t_f}{2\bar{k}}\right).$$

Hence

$$\frac{F_0}{2\bar{k}\left(1 - \frac{t_f}{2\bar{k}}\right)} = \frac{\delta F_0}{\delta t_f} = \frac{\delta F_0}{\delta H} \times \frac{\delta H}{\delta t_f} \quad \ldots\ldots(4)$$

Combining equations (2), (3), and (4), and writing T_f (the applied force per inch width of strip) as $h.t_f$,

$$\frac{\delta H}{\delta T_f} = \frac{F_0}{2c\bar{k}(1 - r)^2\left(1 - \frac{t_f}{2\bar{k}}\right)} \quad \ldots\ldots(5)$$

ECCENTRICITY OF MILL ROLLS

When the rolls of a mill are forced together and rotated, a periodic oscillation usually occurs in the load. This has been recorded in various mills, and a typical variation obtained in the experimental mill is shown in Fig. 20. The wavelength of the disturbance is 31·2 in., and it is clear that the changes are due

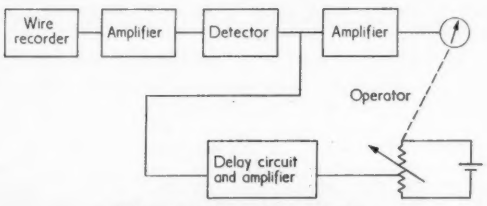


Fig. 19—Training circuit

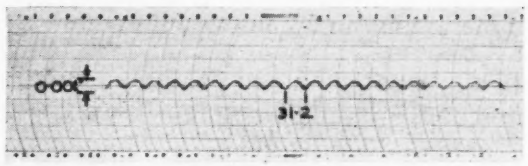


Fig. 20—Variation of mill load in the experimental mill with rolls pressed together

to eccentricity in the rolls, either about their own axes or about the axes of their bearings. The effect on the gauge may be estimated on the F/h diagram by treating the eccentricity as a change in the setting of the rolls, and it may be shown, by an analysis similar to that given on p. 344, that the change in rolled-strip thickness is less than the change in roll setting in a mill that is not fitted with automatic gauge control.

Since the roll eccentricity is equivalent to a change in roll setting, it cannot be corrected by either of the methods of gauge control suggested by Hessenberg and Sims; and, as the frequency of the variation is high at normal rolling speeds, it seems doubtful whether it can be corrected by a simple servo-mechanism. So far, the only way to reduce off-gauge due to eccentricity is to grind the rolls truly round, although in production mills, where the rolls are soon deformed plastically in service, some periodic change in strip thickness along the length of a coil must be expected. Where a roll is supported in hydrodynamically lubricated bearings, the roll barrel must be ground to a common axis with the necks between fixed centres of high precision. This is a difficult operation for large rolls, on the grinders at present available. Where the roll is supported on roller bearings, there is not only the problem of grinding the necks and barrel to a common axis but the bearings themselves may possess eccentricity, which will produce a periodic thickness pattern on the strip. The rolls, therefore, should be ground in their own bearings while preloaded in the direction of the roll separating force. This is a difficult task to carry out in grinding machines of conventional design, and in addition the roller races must subsequently be prevented from turning on the necks after grinding, for the roll barrel will have been ground to an equal and opposite eccentricity to its bearings and necks, and a precession of the races through 180° would double the eccentricity pattern due to the roll-neck bearings alone. It seems that, for the time being, roll eccentricity must be accepted as inevitable in any normal production mill.

THERMAL DISTORTION OF THE MILL COMPONENTS

The energy required to deform a material appears almost entirely as heat in the strip and rolls. The temperature of the mill structure, therefore, rises rapidly when started up from cold on a production run, and thermal expansion takes place in the mill frame, rolls, and roll-neck bearings. The effect is equivalent to a change in the roll setting S_0, and is not corrected by the gauge controller. The change in gauge along the strip rolled in a mill fitted with automatic gauge control will be only slightly greater

than the corresponding change under similar rolling conditions before the control was fitted, and since the controller is quickly reset after each pass the effect of temperature changes is not cumulative. After the mill temperature has reached an equilibrium value the thermal drift in gauge will cease.

CONCLUSIONS

The performance of the prototype closed-loop controllers built on the experimental mill demonstrates that the gauge of rolled strip may be controlled by varying the applied strip tension by an amount depending on an error signal derived from the mill load. Several coils of steel and non-ferrous metals have been rolled consistently to thickness tolerances closer than any previously attained, with apparatus that is suitable for production mills.

Two commonly occurring errors in gauge—due to roll eccentricity and to thermal distortions in the mill—are not corrected by this type of gauge controller. The variations in thickness of the rolled strip from these causes are not large, however, and are well within the accepted tolerances for cold-rolled strip; in any case, these variations appear in all strip rolled at present.

Many problems still remain to be solved in applying the controller to industrial mills; not least are the re-training of mill crews in the use of the device, the rearrangement of rolling schedules, and the abandonment of practices that have been hallowed by time and custom. A more important problem is that of maintenance; the controller is a complex mechanism, and includes several electronic units that will need checking and adjustment by properly trained personnel. These, however, are problems to be faced and overcome in any system of gauge control, if a solution is to be found to the far larger problem of rolling strip at economical speeds to tolerances that conform to the standards and demands of modern engineering.

References

1. W. C. F. Hessenberg and R. B. Sims: *Proc. Inst. Mech. Eng.*, 1952, vol. 166, pp. 75–81.
2. W. C. F. Hessenberg and R. B. Sims: *J. Iron Steel Inst.*, 1951, vol. 168, pp. 155–164.
3. H. Ford: First Report of the Rolling-Mill Research Sub-Committee, *Spec. Rep. No.* 34, pp 69–96: 1946, London, The Iron and Steel Institute.
4. R. B. Sims, J. A. Place, and A. D. Morley: *Engineering*, 1952, vol. 173, pp. 116–119, 137–139.
5. J. Rankine, W. H. Bailey, and F. P. Stanton: *J. Iron Steel Inst.*, 1948, vol. 160, pp. 381–387.
6. R. B. Sims: *Ibid.*, 1952, vol. 172, pp. 415–418.
7. R. B Sims and D. F. Arthur: *Ibid.*, 1952, vol. 172, pp. 285–295.
8. G. E. Stoltz and J. W. Brinks: *Iron Steel Eng.*, 1950, vol. 27, Oct., pp. 69–88.
9. W. E. Hick and J. A. V. Bates: Ministry of Supply Monograph No. 17,204: 1950, London, H.M.S.O.
10. A. Tustin: *Proc. Inst. Elect. Eng.*, 1947, vol. 94, Part 2*A*, pp. 190–202.

Works Trial of the 'T' Method of Automatic Gauge Control

A PROGRESS REPORT

By R. B. Sims, B.Sc., A.M.I.Mech.E., A.Inst..P., J. A. Place, B.Sc., A.Inst.P., and P. R. A. Briggs, B.Sc., A.Inst.P.

SYNOPSIS

Although a complete prototype of the tension method of the B.I.S.R.A. automatic gauge control had been built in the laboratory and had undergone successful trials, it was questioned whether the control would function satisfactorily at high rolling speeds and produce strip of good shape. To resolve these doubts a set of experiments was carried out on the 4-stand 4-high tandem cold mill at the Shotton Works of Messrs. John Summers and Sons Ltd.

755

In a paper on the origin of the thickness variations in rolled strip, Hessenberg and Sims[1] suggested two methods* for the control of gauge in which the error signal is derived from the mill load. In one of these methods, applicable only to cold strip mills, correction is obtained by varying the tensions applied to the strip. This system, which will be termed the 'T' method, has been the subject of laboratory investigation[2] on a pilot-plant scale using the B.I.S.R.A. 10 in. × 10 in. experimental cold rolling mill. Although these experiments have shown clearly that the 'T' method of control can produce strip to a high accuracy in gauge, they left unanswered a number of questions on its behaviour when applied to mills rolling light-gauge wide strip at speeds above 1000 ft./min.

The method of control is intended to correct variations in thickness along the length of the strip, but lateral variations in gauge or reduction, and poor

* Patent applications for these two methods have been lodged in the United Kingdom and abroad.

Paper MW/A/51/52 of the Rolling Committee of the Mechanical Working Division of the British Iron and Steel Research Association, received 10th February, 1953. The views expressed are the authors' and are not necessarily endorsed by the Committee as a body.

shape, occur as a result of changes in the cross-section of the incoming material, variations in lubrication, and thermal changes in the rolls. So complex is the problem that no serious attempt has been made to treat the matter even qualitatively, and it is left to the art of the roller to correct for bad shape as it appears. Variation of strip tension is one way, although by no means the only way, in which the mill operator seeks to produce strip of good shape. One of the best methods of correcting a strip with symmetrically 'riffled' edges is to increase forward tension. If, therefore, a system of gauge correction is installed in which the control of tension is taken away from the roller, there is some danger of strip being of poor shape although to close tolerances in gauge, and in wide strip it is by no means axiomatic that the errors in shape at the cold mill can be removed by the temper mill.

Important doubts in the minds of production personnel arose over the possibility of the mill crew handling this complex mechanism without assistance from trained scientists, and without loss in production of strip. Others doubted the stability of the controller when on a mill rolling at high speeds, as the prototype laboratory experiment was limited to 300 ft./min.

The only way to resolve these doubts was to carry out a long-term investigation on an industrial mill rolling strip of sufficient width and at high speed, at a high rate of production. Such an experiment would also demonstrate the reliability of the equipment and mill instruments, all of which were designed in the B.I.S.R.A. Rolling Laboratory and, at the commencement of the experiment, had not been tried over long periods under works conditions.

The Directors of John Summers and Sons Ltd. made available the 4-stand tandem mill at their works at Shotton for this project and gave every possible assistance and encouragement. The experimental work was commenced in October, 1951, with a survey of the mill, and this account of the progress to date is an attempt to describe the results in a practical way so that mill engineers and production staff may evaluate the possibilities of the method and its application to their own plant.

THE MILL AND THE CONTROL SCHEME

The mill to which the controller was fitted is a Mesta 20 in. and 44 in. × 56 in. 4-high 4-stand tandem installed in 1940. Each stand is powered by a 1500-h.p. motor, and a 250-h.p. motor is fitted to the take-up reel. The motors are Ward–Leonard controlled, with the mill motor armatures supplied from a single generator system. The maximum mill speed is 1700 ft./min., but it is usually run at speeds in the region of 1000–1500 ft./min. The speed of each stand is capable of variation by means of a coarse and vernier rheostat in the motor field. A general view of the mill is shown in Fig. 1, taken from the feed end, and a circuit of one of the mill motors is shown in Fig. 2. The mill normally rolls materials from 24 in. to 45 in. wide and from 0·018 in. to 0·080 in. finished gauge.

The power available at the reel was clearly inadequate for gauge control by forward tension at the 4th stand, and a calculation on the use of back tension

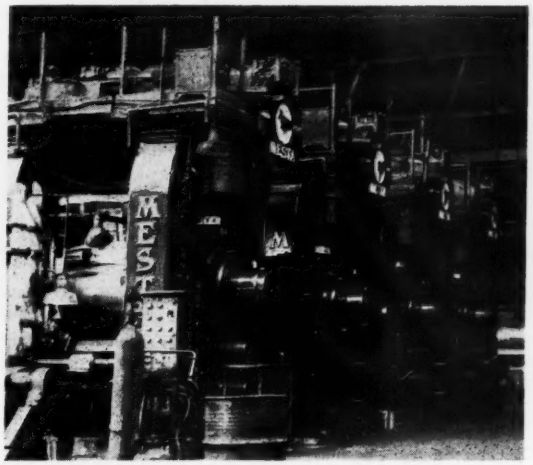


Fig. 1—20 in. and 44 in. × 56 in. 4-high 4-stand tandem mill

for control on this stand indicated that the probability of successful operation was not high. It was decided, therefore, to control the gauge of the strip out of the 3rd stand, using the interstand tension between stands 3 and 4 for control. The reasons for this choice of stand will not be discussed in detail here, except the fact that as the ratio roll-radius/strip-entry thickness (R/H) increases in a stand the power needed for control rises rapidly, and insufficient power was available at the 4th stand motor to effect the substantial reduction in thickness and apply the tension necessary for a control system of sufficient flexibility for the first trials. The control scheme is shown diagrammatically in Fig. 3 and is explained in detail on p. 356.

It will be appreciated that the circuit shown in Fig. 3 cannot achieve the very close control of the strip gauge obtained in the laboratory trials. The interstand tension exerted on strip leaving the 3rd stand will also affect the gauge out of the 4th stand. Hence if the strip entering stand 3 is increasing in thickness, the interstand tension will increase to keep gauge out of stand 3 constant so that the rolled strip thickness leaving stand 4 will decrease in thickness. The strip will be overcorrected, therefore, by an amount which will be, to a first approximation, in advance of the actual error by the distance separating the two stands multiplied by the elongation in the 4th stand. The pass schedules of the mill, particularly

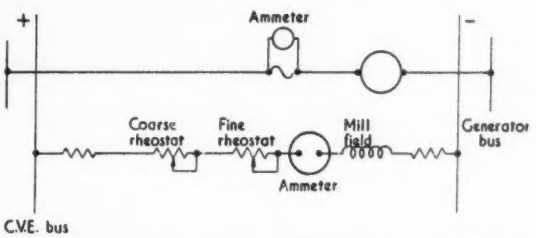


Fig. 2—Mill motor No. 4; field and control before modification

for stands 3 and 4, were arranged to reduce this over-correction to a minimum.

The control system used in the experiments follows the same general pattern as that used on the experimental mill. A detailed description will not be attempted, therefore, since the function of the auxiliary controllers has already been described.[2] Two loadmeters under the screws of stand 3 were connected to measure total load[3] and were supplied with 500 cycles/sec. A.C. from a valve oscillator. The loadmeter output was compared with a reference voltage and the vector difference was made the input of an A.C. amplifier followed by a phase-sensitive demodulator and a push-pull D.C. amplifier. The output stage of the electronic amplifier contained the control windings of an amplidyne connected in the field circuit of the 4th stand motor. Any increase in rolled strip thickness out of stand 3 increased the signal from the loadmeters, and as this signal was compared with a reference which was constant during the pass, the resulting error signal to the amplifier was increased. The amplidyne was arranged to buck the C.V.E. line exciting the mill motor field under these conditions. This weakening of the motor field caused the tension to increase and the gauge error to decrease.

The amplifier was fitted with a servo system for adjusting the reference before the strip was rolled, the controls being interlocked in a way such that the gauge control could not be used until the balancing was complete.

Correction was also applied for the change in thickness of the oil film in the back-up roll bearings. A signal was taken from the tachometer on the mill motor and fed to the D.C. amplifier. It was found that the correction was very close, due mainly to the fact that the diametral clearance in the bearings was of the order of 0·001 in. per inch of diameter, unlike the bearings in the experimental mill.

The controller was fitted with a system of limiting circuits to safeguard the mill and motor. A tensionmeter was fitted between stands 3 and 4, and a signal from it was arranged to saturate the main amplifier when the strip tension decreased below a lower limit (set initially at 4 tons during these preliminary experiments) and when the tension increased above the upper limit, which was adjustable to suit the product being rolled. These limiting circuits also shut down the amplifier in the event of a strip breakage at speed. Another circuit, not used in the experiments described below, was designed to avoid overloading the mill motor. This limiting circuit was placed between the current in the mill motor and the main amplifier, so that, if the current exceeded full load at any time, an auxiliary circuit was arranged to saturate the main amplifier so that no additional signal was possible from the amplidyne.

The coils rolled in this mill are frequently two hot-rolled bands butt-welded together, and there is always a danger of breaking the strip at the junction or marking the rolls when the weld is passed through the mill at speed. It is usual, therefore, to reduce speed when the weld approaches, and, as the speed effect tends to make the strip run thicker at low speeds, the tension in the strip was expected to increase when the mill was fitted with the 'T' method of control. Hence an auxiliary control was fitted which replaced automatic gauge control with automatic control of tension when required, so that if the interstand tension became too great as the mill was slowed down at the approach of the weld, the main gauge control amplifier could be disconnected and

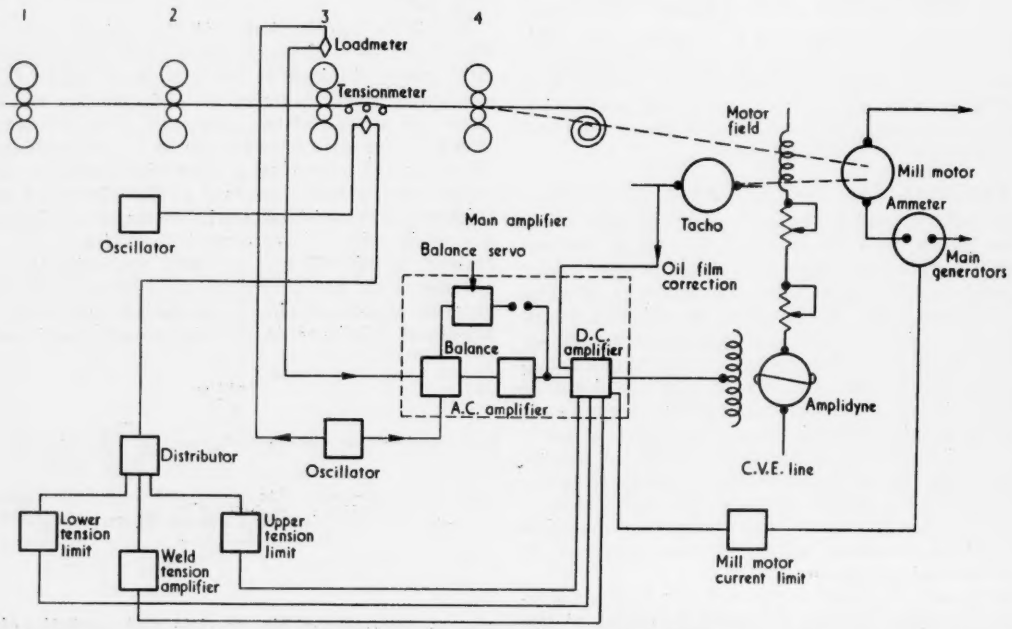


Fig. 3—The gauge and tension control circuit

Fig. 4—Loadmeters fitted in mill

the tension controller, preset at some low value of tension, inserted by depressing a single control stud. On releasing this stud after the passage of the weld, the gauge controller was reconnected in circuit, and the mill could be accelerated again under full automatic gauge control.

The automatic gauge control circuit, the upper and lower tension limits, and the automatic tension control are all independent closed loop systems, each with its own stability problem. They were stabilized separately with *RC* passive networks effecting integration and differentiation of the error signal after the first stage of D.C. amplification. The automatic balancing circuit in the main amplifier was stabilized by feedback between the servo loop elements.

The electronic equipment was connected to the operator's control and to the loadmeter, tensionmeter, and amplidyne, by cabling taken up to a junction box in the crane gantry. In these experiments the apparatus was tested on the bench, but a mobile cabinet had been fabricated to contain the electronic equipment and the stabilized power supplies. The operator's controls were duplicated on the front panel of the cabinet, which also contained meters indicating gauge and tension.

It is not considered good practice to control a D.C. machine in a closed loop control through its field, for at low speeds of rotation of the armature a much larger change is needed in the field current to increase the power output. The system was used in the present experiment because it was impossible to fit a booster of 2000-amp. rating into the armature circuit of the 4th stand motor. Some loss in performance was to be expected, therefore, as the mill speed approached the threading speed of 90 ft./min.

INSTALLATION OF THE LOADMETERS

Two loadmeters, each of 1200 tons capacity, were installed in the mill during the first week of November, 1951. They were constructed to the design described by Sims, Place, and Morley,[3] with the top bearings made to mate with the radiused end of the screws. Two sets of electric resistance strain gauges were fitted, one to indicate load and the other to provide the input to the gauge controller. A stabilized D.C. supply was used to energize the load-indicating

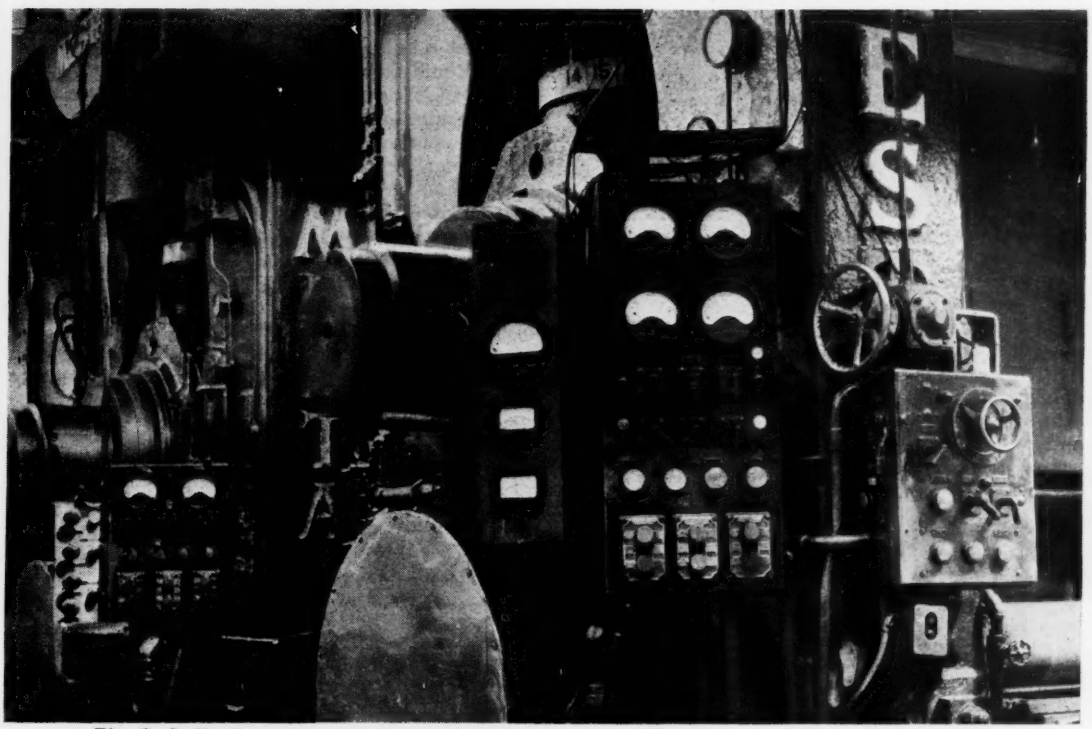


Fig. 5—Indicating meters for load and tension (centre). The gauge control panel is on the left

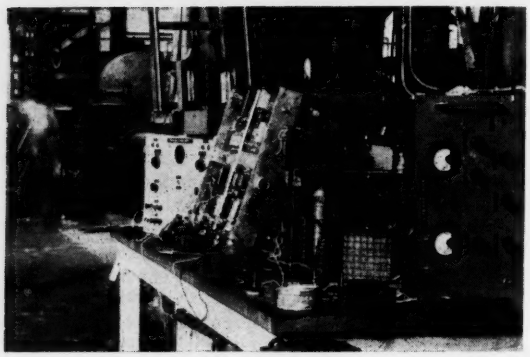


Fig. 6—Bench test

bridges. Indication was made on two microammeters giving full scale deflection for 50 μamp. The installation is shown in Figs. 4 and 5.

These loadmeters have operated without failure since their installation, the only trouble experienced being in the indicating meter circuits.

THE TENSIONMETER

The tensionmeter, which has been described elsewhere,[4] was specially designed for the experiment and was in operation just under seven weeks after the design was begun. Apart from the seizure of a bearing a few minutes after the installation, due to overheating in the labyrinth seal, and the failure of one of the bearings some two months later, it operated without fault from March until August, 1952, when the seals, which were made of pure rubber, were damaged by oil. These have been replaced by temporary seals of a new material.

The strip tension was indicated on a 6-in. meter on the panel attached to the 4th stand and shown in Fig. 5. The tension signal was amplified by a 2-channel A.C. unit designed by Briggs.[3] This amplifier possessed high zero stability and linearity, despite the fact that the capacity of the indicating bridge network was known to vary considerably. It has not required servicing since its installation.

AUTOMATIC TENSION CONTROL

The closed loop control providing automatic tension is a very similar problem in servomechanisms to the full gauge control loop. It was decided, therefore, to examine the stability of the circuit with this form of control before proceeding with gauge control. The tension loop was closed on August 28th, before the amplifiers were assembled in the cabinet. Figure 6 shows the experimental arrangement, and Figs. 7*a* and *b* show records of the tension obtained when rolling a coil immediately before the loop was closed and with the controller in circuit, respectively. It must be emphasized that the variation in tension recorded in Fig. 7*a* is not usual in the mill. In this case it was due to the operator varying the mill speed between wide limits when rolling a particularly poor hot-rolled band which had been imported into the works, and it is shown here as a searching test of performance of the controller. In the pass shown in Fig. 7*b*, with the automatic controller in circuit, the mill speed was similarly varied, but the fluctuations in tension were very much smaller and are apparent because the method of altering tension through the motor field prevented the controller from making full correction during the periods of rapid change in speed. The peak at the beginning of the run was due to the 'snatch' of the strip as it entered the stand, and at the end of the run the small loss of sensitivity as the mill approached threading speed was due to the control being on the motor field. These faults could be removed in a controller and drive which were designed together for a given duty.

AUTOMATIC GAUGE CONTROL

The automatic gauge control loop was closed on August 29th, and several coils approximately 30 in. wide were rolled to thicknesses between 0·036 in. and 0·020 in. Unfortunately, a recording could not be taken from the contact strip micrometer, which in this mill is positioned between the rolls of the fourth stand and a breast roller. The material was measured by hand after trimming and shearing to 6-ft. sheets, and results from two coils finished to 0·020 in. are shown in Fig. 8. Each measurement was made on a separate sheet so that an estimate might be made from them of the accuracy with which the gauge has been controlled along each coil.

In Fig. 8*a* the control was connected into the drive circuit at a finished strip speed of about 400 ft./min. and the mill accelerated to 1400 ft./min. At points marked *A*, *B*, and *C* the mill speed was varied from 1400 ft./min. to about 100 ft./min. At the end of the coil the speed was reduced slowly. It will be seen that the gauge has been held to within ± 0·0015 in. overall, and for the majority of the coil it is to within ± 0·001 in. In particular, the speed effect on slowing down, which is very large on this mill, has been controlled.

In the coil shown in Fig. 8*b* the gauge has been held to ± 0·001 in. after connecting the controller, despite the fact that the strip speed was varied over a wide range. At the point marked *D*, however, a rapid change in gauge occurred which was found in all coils except the one shown in Fig. 8*a*. It occurred during the rapid deceleration of the mill motor under regenerative braking, when the current in the armature falls almost to zero and the influence on gauge of the controller through the motor field is greatly

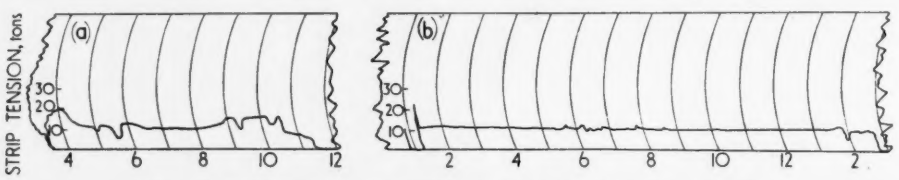


Fig. 7—Strip tension: (*a*) Without control; (*b*) with control

reduced. From the point D to the end of the coil the strip gauge has been held at low strip speeds.

A second set of experiments was made between 17th and 20th September. Unfortunately, a fault occurred in the apparatus which did not permit a set of written records to be obtained. The principal object of these experiments was to test the efficiency of the balancing circuits. The roller was requested to set the mill up to roll to gauge at threading speed, and then the automatic balancing and 'control on' circuits were operated which put the controller into the mill drive circuit. It was found, however, that this procedure took too long, as the roller attempted to change his finished gauge by altering the settings in his first stand. During these experiments the maximum and minimum tension limiting circuits were tested and found to function satisfactorily.

Experiments were carried out on 2nd–3rd October to obtain written records of the gauge variations along coils rolled with automatic control, with the contact micrometer on the mill connected to a recorder/amplifier.*

Coils 28 in. wide for finishing to 0·024 in. thick were used. When rolling the first coil of the batch the mill operator held gauge closely at speed and the amplifier was set by the auxiliary balancing servo. Without changing the screw setting of the 3rd (controlled) stand, the remaining coils were rolled with the controller in circuit. As this work was carried out on a bench test, it was not thought advisable to connect the controller until the initial acceleration period was over.

In Fig. 9*a* a thickness record along the strip is shown when rolling with the controller in circuit. At point Y, the operator raised the screws by a point on the 3rd stand and brought the strip to gauge, thereby decreasing the applied tension.

Figure 9*b* shows the change in gauge on connecting the controller. The strip, which had been running heavy at 0·026 in., was brought to 0·0245 in., the thickness of the preceding coil shown in Fig. 9*a*. At point X the speed was varied rapidly from 1200 ft./min. to threading speed and back to 1200 ft./min., and a change of less than 0·0003 in. occurred in the mean strip thickness.

The thickness variation along a coil rolled with automatic gauge control is shown in Fig. 9*c*. The variation is less than $\pm$ 0·0005 in. about the mean thickness from the time the controller was connected until the contact micrometer was removed at the end of the pass.

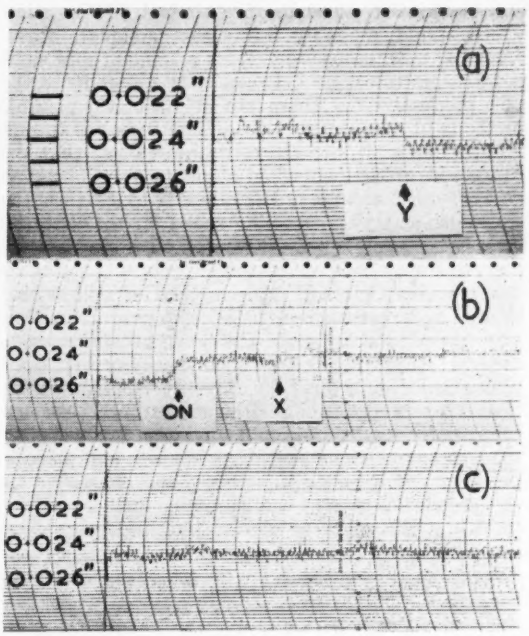


Fig. 9—Thickness variations: (*a*) With controller in circuit; (*b*) change on connecting the controller; (*c*) with automatic gauge control

RESULTS AND CONCLUSIONS REACHED AFTER PRELIMINARY WORK

The results shown on Figs. 7 and 9 indicate clearly that the automatic gauge and tension controls are stable and accurate in operation at speeds of over 1400 ft./min. The gauge of the strip is not controlled to the very close limits obtained on the experimental mill, mainly because the control is on the 3rd stand and not the last stand, and the form of control through the mill motor field is not suitable for high-performance closed-loop regulators, leading to loss of sensitivity at threading speeds and on acceleration and retardation.

There are no indications that the variation in the applied tension caused a poor shape in the final product, although the data on this point are too sparse for any definite conclusion to be reached.

The work has shown clearly the need for intensive development of the control system both operationally

* Manufactured by the General Electric Corporation of America, and loaned by John Summers and Sons, Ltd.

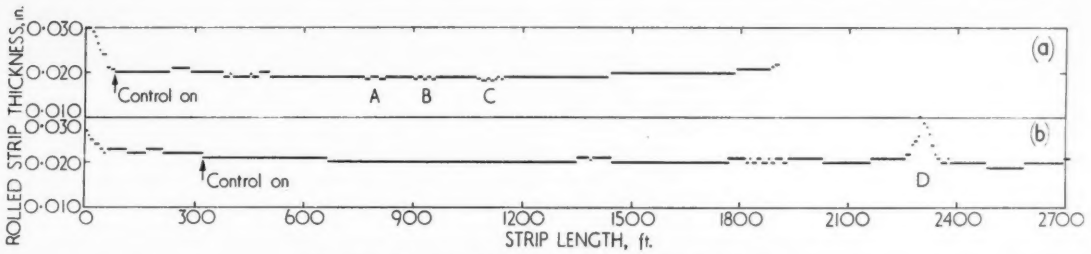


Fig. 8—Measurement of automatic gauge control

and electrically. The balance control, for instance, which would probably have been adequate for a reversing mill, is inadequate for a tandem mill since it increases the time between coils and reduces the mill productivity. There are at least two possible methods of overcoming this difficulty: one is to link a 'gaugemeter'[5] into the control circuit so that the gauge out of the controlled stand may be observed and set with accuracy; the other is to balance the controller at normal running speed on the first coil of a batch and then leave the stand untouched for the remaining coils, as in the experiments made on 2nd and 3rd October. This method of mill setting showed much promise, for little or no production time was wasted; but it would not be so acceptable as a method employing a modified gaugemeter, when rolling orders of a few coils only. It was abundantly clear that the control in its present form was not suitable for use by the mill operator, for there are too few independent safeguards for the mill drive in the event of a maladjustment of the controls. On the electrical side, also, there must be further research to reduce the effect of acceleration and retardation on gauge due to the inertia of the machines and sluggishness in control at low speeds when controlling through the motor field. Finally, since it seems that the method of experimentation will involve a great deal of 'bench' testing, the apparatus built for these experiments is unsuitable, and will require rebuilding to an industrial specification in a rack-mounted system of easily dismountable chassis. In this connection it must be emphasized that, whilst preliminary work of this type can be carried out when the mill is in production, it must be followed by a period when the mill is made wholly available for the experiments.

To sum up, therefore, the 'T' method of gauge control and automatic tension control has worked satisfactorily on a production mill at high speeds without instability developing and without producing strip of poor shape. Strip has been rolled consistently to far closer tolerances than is possible at present in industry.

Acknowledgments

Research on the scale attempted in this project could not have been carried out without the help of all the mill and engineering personnel at the Shotton works of John Summers and Sons Ltd.

Although, unfortunately, it is impossible to acknowledge this assistance individually, the authors would like to thank particularly Mr. A. Reith-Gray for his unfailing support, Mr. H. Edwards for his assistance, Mr. W. H. Corlett and Mr. W. R. Huxley for their organization of the trials on the mill, Mr. S. A. Lewitt and his Electrical Engineering staff who installed much of the equipment, and Mr. R. L. Willott for his help with the tensionmeter. The authors acknowledge with gratitude the assistance of their colleague Mr. W. Bagshaw, and of Mr. H. Ledwood, Mr. J. Middleton, and Mr. J. Holmes and their crews, who handled the mill throughout these trials.

References

1. W. C. F. Hessenberg and R. B. Sims: *Proc. Inst. Mech. Eng.*, 1952, vol. 166, pp. 75–81.
2. R. B. Sims, J. A. Place, and P. R. A. Briggs: *J. Iron Steel Inst.*, 1953, vol. 173, pp. 343–354 (*this issue*).
3. R. B. Sims, J. A. Place, and A. D. Morley: *Engineering*, 1952, vol. 173, pp. 116–119, 137–139.
4. R. B. Sims: *Engineering*, 1952, vol. 174, pp. 262–264.
5. R. B. Sims: *Ibid.*, 1953, vol. 175, pp. 33–35.

Narrow-Side Vertical Cracking in Ten-Ton Ingot Moulds

By A. Jackson, A.Met., F.I.M., and A. N. Whiting, L.I.M.

SYNOPSIS

Trials have been carried out at the Appleby-Frodingham Steel Co. in an attempt to reduce the incidence of premature failure of 10-ton slab moulds by narrow-side vertical cracking. The performance of 100 moulds that had been heated to 750° C. and air-cooled was compared with that of a similar number of untreated moulds used over the same period, and no significant difference was found. To permit rapid stripping, some moulds were lightly dressed at half-life, which in general appears to have a beneficial effect on mould life. The practice of strapping these vertical cracks shortly after they appear is found to be highly successful in prolonging the life of the moulds. 738

The majority of the 10-ton slab moulds used at the Appleby-Frodingham Steel Company are discarded because of a slowly developing horizontal crack about midway up the broad-side, sometimes with a vertical crack growing downwards from the centre of the horizontal crack. The life shows great consistency for this standard type of failure.

There is, however, some tendency towards premature failure caused by vertical cracking through the middle of the narrow-side. This type of crack starts at the bottom of the mould and, on further use, gradually extends up the mould wall until the crack

Paper SM/BA/76–120/52 of the Ingot Moulds Sub-Committee (Ingots Committee) of the Steelmaking Division of the British Iron and Steel Research Association, received 27th November, 1952. The views expressed are the authors', and are not necessarily endorsed by the Committee as a body.

Mr. Jackson is Deputy General Works Manager of the Appleby-Frodingham Steel Company. Mr. Whiting is Assistant Melting-Shop Manager of the same Company.

widens sufficiently to allow the ingot to bleed and cause a 'sticker.'

The new design (Fig. 1) of the 10-ton slab mould previously described* decreases the incidence of narrow-side cracking, but about 10% of the moulds still fail in this way. In an attempt to reduce the incidence still further and to minimize its effect on the mould life, two trials have been carried out:

(1) A batch of moulds was subjected to a stress-relieving annealing treatment and compared with a similar number of untreated moulds

(2) In a number of moulds the vertical crack was strapped in an attempt to prolong the life.

These trials are considered separately below.

TRIAL (1)—ANNEALING METHOD

Effect of Annealing

During trials designed to assess the value of certain variables on the life of 10-ton slab moulds (*loc. cit.*), six annealed moulds were included. The lives of these showed no statistically significant difference that could be attributed to their initial heat treatment, although 'record' lives were obtained from two moulds under different methods of treatment.

The possible effects of annealing have frequently been discussed in technical literature, but no substantial conclusions have been forthcoming. It was therefore decided to conduct a more complete experiment, in an effort to solve this problem.

The moulds are numbered consecutively by the maker, and for the purposes of this trial the odd numbers were selected for the first batch of 50 moulds to be annealed and the even numbers for the second batch. In the whole test, 100 annealed and 100 unannealed moulds were put into service as required; whenever two new moulds were necessary, one treated and one untreated mould were put into service together. They did not remain on adjacent bottom plates but were distributed uniformly throughout the heats, which comprised the general fleet. For many weeks the whole of the 10-ton moulds in use consisted of approximately equal numbers of the two types.

An additional factor in Appleby-Frodingham practice, that of internal dressing, was also studied in this trial. It is known that, to achieve high mould life and economic heating in the soaking pits, the ingot must be stripped a considerable time before the top surface crust is sufficiently strong to withstand appreciable pressure from the stripper ram. This requires that the mould be lifted freely from the ingot; therefore its inner surface must be relatively smooth and free from projections. To achieve this smoothness, it has become customary practice at Appleby-Frodingham to dress the inner surface of the mould by means of a light grinding wheel when the life reaches 75.

Annealing Treatment

The moulds were treated in groups of five in the ingot preheater, each lying on its side, supported on billets, and in such a direction that the furnace gases could pass through the inside and all around the outside, thereby ensuring even heating.

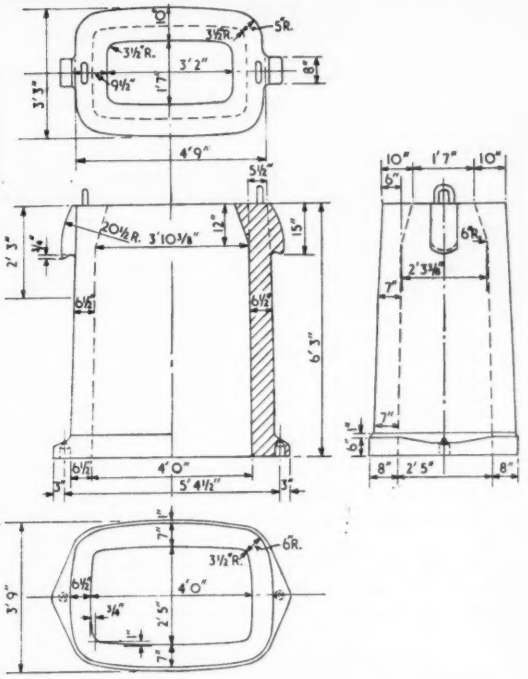


Fig. 1—Design of 10-ton cone-bottom taper-top mould

Prior to charging, the empty furnace was cooled to 800° C. (roof temperature) and five moulds were pushed in on cars. This reduced the temperature to 300°–400° C. immediately, as all five moulds were in the hot zone together.

The roof temperature was raised to 800° C. and maintained for 5 hr., checks by optical pyrometer showing the mould itself to be uniformly heated to 750° C.

The bogies were then drawn, and the moulds standing on the bogies were cooled for about 30–45 min., after which they were removed and cooled normally in air.

Comparison of Annealed with Unannealed Moulds

The mean lives attained in the trial are shown in Table I. The moulds that failed by narrow-side vertical cracking are shown separately, and it will be noted that their lives are considerably less than those of the moulds that failed owing to 'other defects.' In this latter category the majority of the moulds failed by the normal, slowly developing horizontal crack on the broad-side, and the remainder failed by wear of the inner surface, which rendered stripping unduly difficult and seriously marked the ingot. During the 12-month trial period, more than 320,000 tons of ingots were made in the test moulds.

It is obvious from Table I that this method of heat treatment of the moulds has effected no significant change in the mould life, nor has it decreased the incidence of narrow-side vertical cracking. In view of these results and of the heating and handling costs

* A. JACKSON: *J. Iron Steel Inst.*, 1951, vol. 167, pp. 289–301.

Table I

LIVES OF ANNEALED AND UNANNEALED MOULDS

Moulds	Narrow-Side Vertical Cracking		Other De ec	
	No. of Moulds	Average Life	No. of Moulds	Average Life
Annealed:				
Dressed	6	107	44	174
Not dressed	8	98	42	164
	14	102	86	169
Unannealed:				
Dressed	4	115	46	167
Not dressed	7	96	43	167
	11	97	89	167

involved, the practice of annealing the moulds prior to use has now been discontinued.

There is an indication, however, that the practice of dressing the inner surface of the mould, which is carried out essentially to maintain easy stripping, gives a small general increase in mould life.

TRIAL (2)—STRAPPING METHOD

In a further attempt to minimize the effect of narrow-side vertical cracking on mould life, it was decided to strap the crack by securing a steel strip (12 in. × 4 in. × 1 in.) across it by two bolts ($1\frac{1}{2}$-in. dia.) that were screwed 3 in. into the mould wall. Figure 2 shows a crack that was strapped very shortly after appearing on the mould wall.

The first trial of this method was conducted on 20 of the ordinary 10-ton moulds shown in Fig. 1, and the results are shown in the following table.

Fig. 2—Crack on mould wall, strapped very shortly after appearance

Ordinary 10-ton Mould

		No. of Moulds	Average Life
Moulds failing by defects other than narrow-side vertical cracking (32-weeks period)		123	180
Moulds failing by narrow-side vertical cracking (same period)		14	108
20 strapped moulds	Average life at which strapped		114·8
	Average life at which discarded		159·3
	Average increase in life		44·5

Thus, by strapping the 20 moulds in which narrow-side vertical cracks had appeared, a total number of 890 extra lives was obtained from them.

The success of this trial led to a similar trial on the brick-top 10-ton mould. This mould has dimensions that are identical with those of the ordinary 10-ton mould (as in Fig. 1), together with a recess for the brick top in the top 15 in. of the mould. The following results were obtained in this case:

Brick-Top 10-ton Moulds

		No. of Moulds	Average Life
Moulds failing by defects other than narrow-side vertical cracking (32-weeks period)		10	75
Moulds failing by narrow-side vertical cracking (same period)		10	33
13 strapped moulds	Average life at which strapped		29·6
	Average life at which discarded		46·2
	Average increase in life		16·6

In this case a total number of 216 extra lives was obtained by strapping the 13 moulds.

The method is thus very successful in minimizing the harmful effect of narrow-side vertical cracking. To achieve the greatest success, the crack should be strapped as soon as possible after appearing and before it has progressed more than about 1 ft. up the mould wall. By this means, the mould life can be appreciably prolonged, and the mould is eventually discarded either because the normal broad-side has cracked or because the inner walls have become worn.

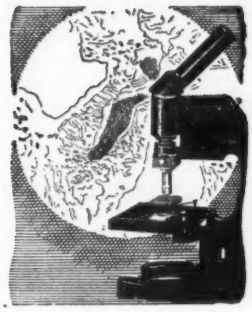
CONCLUSIONS

Annealing of the 10-ton moulds for 5 hr. at 750° C., followed by air-cooling, had no significant effect on mould life nor did it bring about any reduction in the number of early failures caused by vertical cracks on the narrow-side. This expensive practice has therefore been discontinued.

Dressing of the inner faces of the moulds, which is carried out essentially to maintain easy stripping, appears to result in a small general increase in mould life. It is customary at Appleby-Frodingham to dress the inner surfaces of each mould when it reaches a life of 75.

The practice of strapping narrow-side vertical cracks appreciably reduces their harmful effect on the mould life. On both the ordinary type and the brick-top type of 10-ton mould, the life of moulds developing this type of defect can be prolonged by about 40–50%, provided that the strap is secured very shortly after the crack appears.

Carbides in Chromium, Molybdenum, and Tungsten Steels



By Kehsin Kuo

SYNOPSIS

The occurrence, nature, and internal relations of carbides in chromium, molybdenum, and tungsten steels with respect to the steel composition and heat-treatment have been studied by electrolytic extraction and X-ray powder methods.

In chromium steels the carbide reaction $Fe_3C \rightarrow Cr_7C_3$ takes place in two stages : the concentration of chromium in Fe_3C and the transformation of Fe_3C into Cr_7C_3. Carbide Fe_3C occurs also in tempered high-chromium steels ; later it passes through the state Cr_7C_3 and reaches the final stable state $Cr_{23}C_6$. An equilibrium diagram below the A_1 temperature, based upon α-iron, Fe_3C, Cr_7C_3, and $Cr_{23}C_6$, is constructed.

In the system Fe–W–C, the stable carbides are Fe_3C, WC, and M_6C ; Fe_3C appears as a stable carbide only when the atomic ratio W : C of the steel is smaller than unity, otherwise M_6C appears. Based upon these carbides, an equilibrium diagram below A_1 is drawn. Carbides $M_{23}C_6$ and W_2C are metastable. The six different carbide reactions observed are explained in the light of the compositions of the matrix and carbides.

Two more carbides exist in the system Fe–Mo–C than in the system Fe–W–C ; these are a low-temperature modification of MoC and a third Fe–Mo double carbide M_aC_b, and both are metastable carbides. The carbide reactions and the stable carbides in this system are similar to those in the system Fe–W–C. Carbide MoC with WC structure is found in many steels after long tempering.

The carbide reactions in these three systems are compared and the different mechanisms are discussed. 719

Introduction

Among the common alloy elements added to steels, the chromium group of elements (chromium, molybdenum, and tungsten) plays a unique and important rôle, not only because these are the main alloy elements in tool and heat-resisting steels but also because of the complex constitution exhibited by steels containing these elements.

The transition elements on the left side of the chromium group of elements in the periodic table have such a high affinity for carbon that they form with carbon the NaCl-type carbide. Only when there is carbon in excess after the formation of this NaCl-type carbide will cementite be present in the steel. Although the elements on the right side of the chromium group of elements (excluding iron) form their own carbides with carbon, in steels they will be dissolved in the cementite without forming any new phase.

It is shown in the following pages that chromium introduces two, molybdenum five, and tungsten four new carbides into steels. All except two of these carbides were known before, chiefly owing to Westgren and Phragmén, but not all of them have been found in steels. However, no systematic work has been done to elucidate the nature and occurrence of these carbides and their internal relationships in respect of chemical composition and heat-treatment of the steel, and consequently wrong conclusions and speculations are often found in current literature. Owing to the theoretical interest and the technical importance of these steels, investigations on pure Fe–Cr–C, Fe–Mo–C, and Fe–W–C systems were undertaken.

The investigation had two main objects: (i) to determine the equilibrium diagrams of these three systems below the A_1 temperature, and (ii) to study the carbide reactions, or carbide phase changes, in the isothermally transformed and quenched-and-tempered states approaching equilibrium. To eliminate the influence of foreign elements, pure Fe–M–C ternary alloys were made in a laboratory-scale high-frequency furnace. Except in special cases, carbides were completely dissolved at the austenitizing temperature, so that there was no complication of undissolved carbide. The carbide reactions have been critically examined and their special features have been pointed out. A tentative but consistent explanation of the observed phenomena has also been given.

The X-ray powder method has been used as a means of identifying the different carbides, since it gives definite evidence of the presence of any carbide. The accuracy of identification has been greatly improved by concentrating the carbide phase by dissolving the iron phase anodically in an acid electro-

Manuscript first received 29th August, 1952, and in its final form 10th November, 1952.

This project is a part of the technical research programme sponsored by Jernkontoret (The Swedish Ironmasters' Association), and was led by a Committee whose members are H. Nathorst (Chairman), A. Hultgren, G. Hägg, and K. Kuo.

Mr. Kuo is at the Institute of Chemistry, University of Uppsala, Sweden.

lyte. In addition to X-rays, chemical analysis and microscopic examination have also been used in this investigation. The experimental details are given in the Appendix. No attempt has been made to review all the literature concerning the carbides occurring in these systems, since excellent reviews already exist[1, 2]; only those pertaining to this paper will be briefly discussed.

Section I—CARBIDES IN CHROMIUM STEELS

EQUILIBRIUM DIAGRAM OF THE SYSTEM Fe–Cr–C BELOW A_1

Carbides in the System Fe–Cr–C

Chromium forms three carbides with carbon, and their crystallographic data, determined by Westgren and his associates (*see* references 1 and 2), are given in Table I. There is no double carbide in this system, but the solubilities of iron in chromium carbides and of chromium in cementite are fairly high, as is shown in the last column of Table I. With the exception of Cr_3C_2, all chromium carbides have been found in chromium steels.

Goldschmidt[3] reported a new carbide in chromium steels with an NaCl-type lattice, $a = 3{\cdot}618$Å. Lashko and Nesterova[4] found the same phase, which they called the metastable γ' phase, in the early stage of tempering of high-chromium steels. The present author has made the same observation, but the phase in question is considered to be the retained austenite undissolved in the electrolyte, rather than a new carbide.

Jack[5] found the ε-iron carbide as a low-temperature transition phase of tempering martensite, and it is to be expected that this carbide will also appear in chromium steels tempered at low temperatures. However, since this investigation deals only with carbides present in steels at 400° C. and above, this low-temperature carbide has not been encountered.

Concentration of Chromium in Fe_3C

It is a well-known fact that chromium in steels concentrates itself in cementite in preference to ferrite or austenite. The diffusivity of carbon is about 10^6 times as high as that of iron or chromium. When cementite is formed at low temperatures, only the carbon atoms migrate, leaving the Cr : Fe proportion in the carbide unchanged. Therefore, the cementite, formed first, will inherit its Cr : Fe ratio from the matrix. However, if the steel specimen is kept at a temperature such that the metal atoms also gain enough mobility, chromium will migrate from the matrix into cementite.* As is shown in Table II, the partition ratio of chromium between cementite and ferrite at equilibrium is 28 : 1.

Equilibrium Diagram

The composition and heat-treatment of the steels studied are shown in Table II. As already pointed out, both the concentration of chromium in cementite and the carbide transformation require the diffusion of chromium, and therefore the steels were either tempered or annealed at 700° C. for long periods. After 120 hr. at this temperature, equilibrium was attained. The chemical and X-ray analyses of the isolated carbide are also shown in Table II. The composition of ferrite was obtained by analysing the electrolyte in which the steel specimen had been dissolved anodically. From these data the equilibrium diagram of the Fe–Cr–C system at 700° C. was drawn (Fig. 1). This diagram agrees well with the findings of Heczko[7] as modified by Dannöhl,[8] but not with the diagrams constructed by Westgren and Phragmén[9] and Tofaute, Küttner, and Büttinghaus.[10] Westgren and Phragmén's specimens, from which carbides were isolated and analysed, were not annealed, and hence the equilibrium state has not been attained. Tofaute and his co-workers assumed that Cr_7C_3 and Fe_3C form a pseudo-binary system, which is inconsistent with the different formulae of these two carbides.

From Fig. 1 it can be seen that Fe_3C can dissolve about 18% of chromium, whilst Cr_7C_3 and $Cr_{23}C_6$ can

* Hultgren[6] called the cementite with inherited metal composition para-cementite and the cementite with the alloy element redistributed ortho-cementite.

Table I

CARBIDES IN THE SYSTEM Fe–Cr–C

Carbide	Space Lattice	Lattice Dimensions, Å.	No. of Formula-Units per Unit Cell	Remarks
Fe_3C*	Orthorhombic, simple	$a = 4{\cdot}514$ $b = 5{\cdot}079$ $c = 6{\cdot}730$	4	Chromium can replace iron up to 18%
$Cr_{23}C_6$	Cubic, face-centred	$a = 10{\cdot}64$	4	Iron can replace chromium up to 35%
Cr_7C_3	Trigonal	$a = 13{\cdot}98$ $c = 4{\cdot}523$	8	Iron can replace chromium in equilibrium with α-iron from 30 to 50%
Cr_3C_2	Orthorhombic, simple	$a = 2{\cdot}821$ $b = 5{\cdot}52$ $c = 11{\cdot}46$	4	Not identified in chromium steels

* The formulae, except those specially referred to, merely denote the type of carbide and the metal : carbon proportion, leaving the composition of the metal portion of the carbide unspecified. In the case of double carbides, M is used to represent the metals (*e.g.*, $Fe_{21}W_2C_6$ is written $M_{23}C_6$); $Cr_{23}C_6$ and $M_{23}C_6$ are not the same carbide

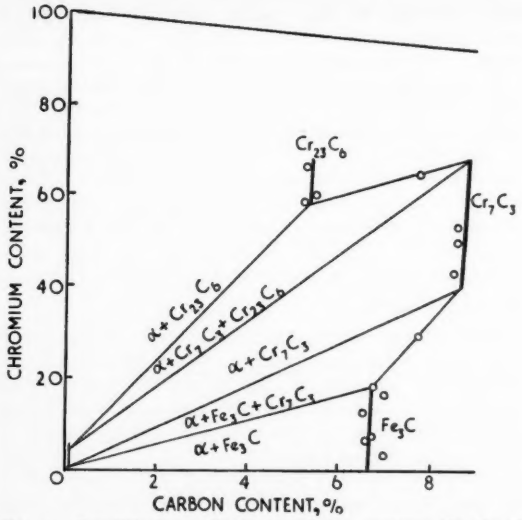


Fig. 1—Equilibrium diagram of the system Fe–Cr–C at 700° C.

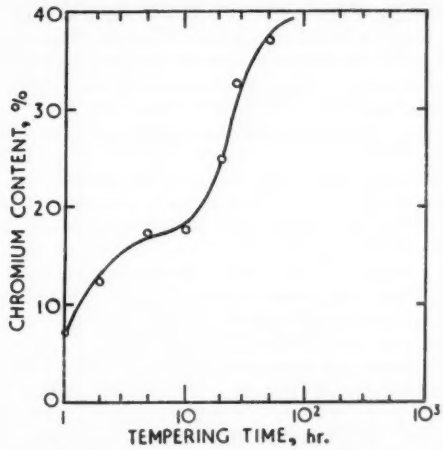


Fig. 2—Carbide reaction $Fe_3C \rightarrow Cr_7C_3$ in steel *C*8 (0·51%C, 3·13%Cr) tempered at 560° C.

dissolve as much as 50% and 35% of iron, respectively. The ferrite in equilibrium with both Cr_7C_3 and $Cr_{23}C_6$ contains 5·6% of chromium.

CARBIDE REACTIONS IN CHROMIUM STEELS

Carbide Reaction $Fe_3C \rightarrow Cr_7C_3$

Steel *C*8 (0·51% C, 3·13% Cr) was chosen for the study of this reaction. After equilibrium is reached the carbide in this steel is Cr_7C_3, containing 41·0% of chromium. Steels with similar compositions have also been studied by Lyman and Troiano,[11] Bokshtein,[12] and Wever and Koch.[13]

The steel specimens were quenched from 1300° C. and were then tempered for 1 hr. at different temperatures or at 560° C. for different periods. It was found that after tempering for 1 hr. the reaction $Fe_3C \rightarrow Cr_7C_3$ takes place between 580° and 620° C. At 560° C. this reaction was noticed after 10 hr. and completed after 20 hr. tempering. In the constant-temperature tempering series, both X-ray and chemical analyses of the carbide were made, the results of which are shown in Fig. 2. Similar curves were obtained by Bokshtein[12] and Wever and Koch.[13]

The conclusion drawn from this is that the carbide reaction $Fe_3C \rightarrow Cr_7C_3$ takes place in two stages. The first stage is the concentration of chromium in Fe_3C, which proceeds until the chromium content has reached about 20%, *i.e.*, the saturation value. The second stage is the replacement of Fe_3C by Cr_7C_3, accompanied by an abrupt increase in chromium and a decrease in the amount of carbide. Lyman and Troiano's findings[11] that, in a similar steel, the carbide reaction $Fe_3C \rightarrow Cr_7C_3$ takes place faster in fine pearlite than in coarse pearlite, while in the latter case the diffusion path in the cementite is greater, may serve to prove that the build-up of chromium in Fe_3C is the rate-determining process.

Table II

EQUILIBRIUM STATES OF THE SYSTEM Fe–Cr–C AT 700° C.

Steel				Carbide Residue Extracted from Steel Specimen			Ferrite (Electrolyte) Cr, %	Partition Ratio of Cr Between Fe_3C and α-Fe
No.	C, %	Cr, %	Heat-Treatment*	X-Ray Analysis	C, %	Cr, %		
C 1	1·02	0·48	*a*	Fe_3C	7·00	3·10	0·12	25·6:1
C 2	0·98	0·99	*a*	Fe_3C	6·67	6·32	...	...
C 3	0·48	0·53	*a*	Fe_3C	6·74	6·54	0·23	28·4:1
C 4	0·49	1·05	*a*	Fe_3C	6·46	10·9	0·41	26·6:1
C 5	0·51	1·55	*a*	Fe_3C	6·98	16·2	0·59	27·5:1
C 6	0·97	2·91	*a*	Fe_3C	6·77	18·1	0·59	30·7:1
C 7	1·00	4·09	*a*	$Fe_3C + Cr_7C_3$	7·70	27·7	0·86	...
C 8	0·51	3·13	*a*	Cr_7C_3	8·50	41·0	0·93	...
C 9	2·05	11·2	*b*	Cr_7C_3	8·56	47·0	1·23	...
C10	0·19	2·91	*a*	Cr_7C_3	8·60	50·0	1·56	...
C11	0·94	12·2	*b*	$Cr_7C_3 + Cr_{23}C_6$	7·95	64·4	5·57	...
C12	0·39	12·4	*b*	$Cr_{23}C_6$	5·50	59·6	...	...
C13	0·30	12·2	*b*	$Cr_{23}C_6$	5·64	60·6	8·58	...
C14	0·22	24·2	*b*	$Cr_{23}C_6$	5·53	64·6	22·8	...

* *a* = Quenched from 1300° C. and tempered at 700° C. for 120 hr.
b = As-received, then annealed at 700° C. for 250 hr.

Balluffi, Cohen, and Averbach[14] suggest that the Fe_3C does not transform into Cr_7C_3 *in situ* but actually dissolves in the ferrite matrix while Cr_7C_3 precipitates grow elsewhere. Their evidence is based on electron photomicrographs, which show that the Fe_3C has a plate-like structure whereas Cr_7C_3 is spheroidal in form. There seems no doubt that the Fe_3C formed at low temperature has a plate-like appearance, but it can also have spheroidal form if allowance is made for coagulation and growth. In any case, the evidence is not convincing enough to close the discussion.

If Cr_7C_3 were formed from the ferrite, as suggested by Balluffi and his co-workers,[14] the probability of forming this chromium-rich carbide should be greater in a matrix with higher chromium content. After the concentration of chromium in Fe_3C, the chromium content of the matrix is considerably reduced and precipitation of Cr_7C_3 should be rendered more difficult. The experimental results, however, show the reverse and make this mechanism less likely.

From these arguments it is tentatively concluded that Cr_7C_3 is probably formed by the reaction of Fe_3C with the matrix.

Carbide Reaction $Cr_7C_3 \rightarrow Cr_{23}C_6$

For this purpose, steel *C*13 (0·30% C, 12·2% Cr) was studied. After 30 min. at 1100° C., all carbides were dissolved in the austenite. The steel specimens were then quenched and were tempered for 1 hr. from 400° to 750° C. at 25° intervals. X-ray analyses of the carbides isolated from the steel specimens show the following results: below 500° C., only Fe_3C; 525–625° C., only Cr_7C_3; 650–675° C., Cr_7C_3 and $Cr_{23}C_6$; 700° C. and above, only $Cr_{23}C_6$.

It was thus proved that Fe_3C occurs also in tempered high-chromium steels, a phenomenon not definitely established before. The mechanism of the carbide reaction $Cr_7C_3 \rightarrow Cr_{23}C_6$ has not been studied, but it may follow the same steps as outlined above for the reaction $Fe_3C \rightarrow Cr_7C_3$. An increase of the chromium content in Cr_7C_3 also occurs, as shown by steels *C*8, *C*9, and *C*10 in Table II. The chromium content increases from 41% to 50%.

Resolution of $Cr_{23}C_6$

Steel *C*11 (0·94% C, 12·2% Cr) contains Cr_7C_3 and $Cr_{23}C_6$ after long annealing. When this steel was austenitized at 1100° C., the $Cr_{23}C_6$ was found to be completely dissolved, leaving Cr_7C_3 as the only undissolved carbide. This has been found by Lane and Grant[15] in high-temperature alloys, and was also reported in 1936 by Tofaute and his co-workers.[10] In many of their equilibrium diagrams, at high temperatures the $Cr_7C_3 + Cr_{23}C_6$ two-phase region is replaced by a Cr_7C_3 single-phase region. No proper explanation has yet been given for this phenomenon. By comparison with the chromium–carbon binary system, Lane and Grant[15] postulated that perhaps this is due to the low melting point of $Cr_{23}C_6$ (1518° C.) compared with that of Cr_7C_3 (1782° C.). Lacking a better explanation, this postulate is accepted but is not considered to be final.

Section II—CARBIDES IN TUNGSTEN STEELS*

EQUILIBRIUM DIAGRAM OF THE SYSTEM Fe–W–C BELOW A_1

Carbides in the System Fe-W-C

Tungsten forms two carbides with carbon and two double carbides with iron and carbon, and all these carbides have been identified in tungsten steels; their crystallographic data are given in Table III. The two double carbides have rather complicated structures and their compositions vary within broad limits. For this reason they will be denoted by the general formulae M_6C and $M_{23}C_6$, instead of the formulae Fe_3W_3C (or Fe_4W_2C) and $Fe_{21}W_2C_6$ used by Westgren and Phragmén.

Wood and Wainwright[16] found a transition compound, which was later replaced by M_6C after prolonged heating, in a tungsten magnet steel heated for a few minutes at 850° C. As is shown on p. 369, this transition phase is probably the W_2C carbide. Goldschmidt[17] recently found that, during the tempering of high-speed steels between 620° and 800° C., a fine dispersion of M_6C forms first which, after passing through a "transition phase *T*," changes into $M_{23}C_6$. This transition phase, according to Goldschmidt, does not correspond to any known phase in the Fe–W–C system. It has not been found by the present author.

Modifications of M_6C have been reported,[17] but they have not been found in tungsten steels.

Isothermally Transformed Steel Specimens

After austenitizing at 1300° C., the steel specimens were transformed isothermally at 700° C. Even in high-tungsten steels the transformation was completed within 10 hr. The results of carbide phase analysis are shown in Table IV. The steels are arranged in order of increasing atomic ratio of the tungsten to the carbon content of the steel, since both tungsten and carbon take part in the formation of carbides. If the discussion is limited to the carbides present immediately after complete austenite transformation, the results obtained in this investigation agree well, in principle, with the findings of Krainer[18] and Rosenbaum.[19] As the atomic ratio W: C of the steel increases, the carbide found is Fe_3C, $M_{23}C_6$, or M_6C. In steel *W*7 the transformation products are α-iron plus $M_{23}C_6$ and M_6C, but in a partially transformed specimen (10% transformation) no $M_{23}C_6$ exists. This steel has an atomic ratio W: C of 0·67, whilst M_6C (with the composition Fe_4W_2C) has a ratio of 2·0. The fact that M_6C is formed in this steel can perhaps be explained by the temporary concentration of tungsten atoms caused by fluctuations in the composition. After the formation of some M_6C, the tungsten content of the matrix (untransformed austenite) drops materially, and this may account for the formation of $M_{23}C_6$ in the later stage of austenite transformation.

No evidence has been found that Fe_3C forms before $M_{23}C_6$ and M_6C. It seems that these two iron–

* The essential material presented in this section has been published, in Swedish, in *Jernkontorets Ann.*, 1952, vol. 136, pp. 156–168.

Table III
CARBIDES IN THE SYSTEM Fe–W–C

Carbide	Space Lattice	Lattice Dimensions, Å.	No. of Formula-Units per Unit Cell	Remarks
Fe_3C	*See* Table I			
WC	Hexagonal	$a = 2{\cdot}901$ $c = 2{\cdot}830$	1	Wood[38] identified WC in a 6% tungsten magnet steel, and Takeda[23] found this carbide in many tungsten steels
W_2C*	Hexagonal	$a = 2{\cdot}986$ $c = 4{\cdot}712$	1	Arkharov, Kvater, and Kiselev[39] found W_2C in a burnt high-speed steel with high carbon content (1·43%) and low tungsten content (11·26%)
$M_{23}C_6$	Cubic, face-centred	$a = 10{\cdot}52$	4	First found by Morral, Phragmén, and Westgren[40] in a 6% tungsten magnet steel. This carbide is isomorphous with $Cr_{23}C_6$ and has an ideal composition $Fe_{21}W_2C_6$
M_6C	Cubic, face-centred	$a = 11{\cdot}08$	16	First found by Westgren and Lindh[41] in high-speed steels, and later also by Wood[38] in tungsten magnet steels. Its composition limits are Fe_4W_2C and Fe_3W_3C

* Becker[42] has reported a high-temperature modification of W_2C, but this has not been definitely established. Recently, Lander and Germer[43] found a cubic modification of this carbide with NaCl-type structure; this also needs to be confirmed[44]

tungsten double carbides form directly from the austenite at 700° C. At low transformation temperatures, Fe_3C occurred together with $M_{23}C_6$, as in the case of steel *W*3 when it was transformed at 600° C. At still lower temperatures, *e.g.*, 500° C., Fe_3C occurred alone, because at this low temperature iron and tungsten atoms lose their mobility to a great extent, whereas carbon atoms are still highly mobile, owing to their small atomic size.

Carbides W_2C and WC have not been found in the isothermally transformed steel specimens just after the completion of transformation.

Metastable Equilibrium Diagram

If the isothermally transformed steel specimens are kept at 700° C. for longer periods, carbide reactions will take place. Carbide WC appeared in all the steels after annealing for 50–100 hr., and the relative proportions of the carbides changed. However, the carbide reactions in tungsten steels will be considered at greater length in connection with the quenched and tempered specimens.

As is well known, a suitable amount of chromium prohibits the formation of WC in tungsten tool steels. Therefore, a metastable equilibrium diagram at 700° C. based upon α-iron, Fe_3C, $M_{23}C_6$, and M_6C can be drawn as shown in Fig. 3. A few Fe–W–Cr–C alloys have also been studied, and it was found that chromium, forming $Cr_{23}C_6$ with carbon, widens the steel composition range within which $M_{23}C_6$ will appear. However, its effect on tungsten steels is much less than on molybdenum steels. That is, chromium

Table IV
EQUILIBRIUM STATES OF THE SYSTEM Fe–W–C AT 700° C.

Steel				X-Ray Analysis of Carbide Residue Extracted from Steel Specimen	
No.	C, %	W, %	Atomic Ratio W:C	After Isothermal Transformation* for 10 hr. at 700° C. Metastable Equilibrium	After Annealing* for 2000 hr. at 700° C. Stable Equilibrium
W 1	0·98	0·49	0·033	Fe_3C	$Fe_3C + WC$
W 2	1·16	1·16	0·065	Fe_3C	$Fe_3C + WC$
W 3	0·65	1·50	0·15	$M_{23}C_6$	$Fe_3C + WC$
W 4	0·89	2·62	0·19	$M_{23}C_6$	$Fe_3C + WC$
W 5	0·55	1·96	0·23	$M_{23}C_6 + M_6C$	$Fe_3C + WC$
W 6	0·59	3·62	0·40	$M_{23}C_6 + M_6C$	$WC + Fe_3C$
W 7	0·60	6·12	0·67	$M_{23}C_6 + M_6C$	$WC + Fe_3C$
W 8	0·27	5·45	1·32	$M_6C + M_{23}C_6$	$WC + M_6C$
W 9	0·34	8·15	1·57	$M_6C + M_{23}C_6$	$WC + M_6C$
*W*10	0·32	9·56	1·96	M_6C	$M_6C + WC$

* Austenitizing temperature: 1300° C.

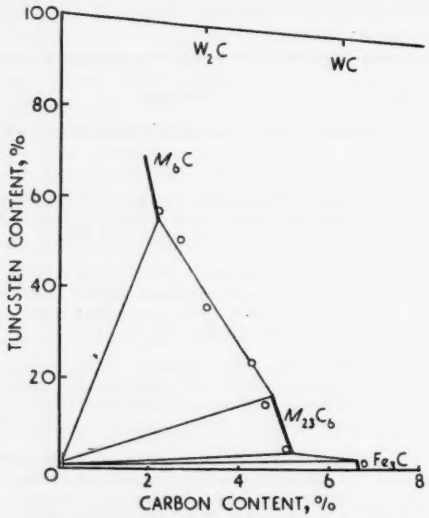


Fig. 3—Metastable equilibrium diagram of the system Fe–W–C at 700° C.

modifies this diagram only slightly, and the latter may therefore be applied equally well to tungsten tool steels with low chromium contents.

Carbide $M_{23}C_6$ may contain 5–15% of tungsten, *i.e.*, less than the proportion corresponding to the formula $Fe_{21}W_2C_6$ (22·8% W). The lower limit of the tungsten content of the carbide M_6C is about 55%, *i.e.*, somewhat lower than that given by the formula Fe_4W_2C (61·2% W).

Stable Equilibrium Diagram

After annealing at 700° C. for 2000 hr., equilibrium was reached; the carbides found in equilibrium with α-iron at this temperature are shown in the last column of Table IV. It can easily be seen that if the atomic ratio W: C is smaller than unity, Fe_3C and WC co-exist as stable carbides; otherwise, M_6C and WC co-exist as stable carbides. With this ratio at unity, WC should exist alone (if the small amount of tungsten and carbon dissolved in ferrite is disregarded). Carbide WC may not exist in steel with very low tungsten content, but this possibility has not been investigated.

The lattice dimensions of WC extracted from tungsten steels are $a = 2·832$Å., $c = 2·903$Å. (*i.e.*, the same as for pure WC). This means that the solubility of iron in WC is very limited. The findings of Adelsköld, Sundelin, and Westgren[20] for the solubility of nickel or cobalt in WC were similar. The solubility of tungsten in Fe_3C is not known exactly. In steel *W*1 the carbide mixture, Fe_3C and WC, was found to contain about 1·4% of tungsten, so that the amount of tungsten dissolved in Fe_3C cannot be higher than 1%. Kuo and Hultgren[21] found that in a 0·49% C, 0·45% W steel, tempered at 700° C. for 24 hr., the carbide residue contains 0·90% W. X-ray analysis shows the existence of only Fe_3C, but a small amount of tungsten-containing carbide may exist.

The stable equilibrium diagram shown in Fig. 4 is based upon these findings. Microscopic examination did not show any sign of graphitization in these tungsten steels after annealing for 2000 hr. at 700° C., but this does not rule out the possibility of graphitization after still longer annealing.

The similarity between the systems Fe–W–C and Co–W–C is obvious. Only WC can be found in sintered carbide made of WC and 6% of cobalt. However, as was shown by Berg and Krainer,[22] M_6C will appear in these alloys if there is some decarburization during sintering, and graphite will occur if there is carbon in excess, since Co_3C is not stable at high temperatures.

Stability of M_6C

Takeda's postulate[23] that M_6C is metastable with respect to WC in the same way that Fe_3C is metastable with respect to graphite was not substantiated by this investigation. In special cases, *e.g.*, in steels *W*5, *W*6, and *W*7 (*see* Table IV) which have an atomic ratio W: C smaller than unity, M_6C does disappear after long annealing and is replaced by WC. However, M_6C will remain together with WC as a stable carbide if the atomic ratio is larger than unity. Sandford and Trent[24] also have doubted Takeda's conclusion. Their opinion about the η phase (Co_3W_3C) in sintered carbide is, in certain respects, similar to the present author's view concerning M_6C in tungsten steels.

Carbide Fe_3W_3C has been synthesized from WC + 2W + 3Fe at 1800° C. After the reaction was completed there was no trace of WC remaining, and the Fe_3W_3C thus produced was stable even after long heating at low temperatures. Thus it is clear that M_6C would exist alone if there were enough tungsten to form Fe_3W_3C, *i.e.*, if the atomic ratio W: C of the alloy were larger than 3.

Carbide M_6C is a common double carbide of transition metals. Metal pairs known to form this carbide are W–Fe, W–Co, W–Ni, Mo–Fe, and Mo–Co. Recently, Binder[25] reported the M_6C phase in

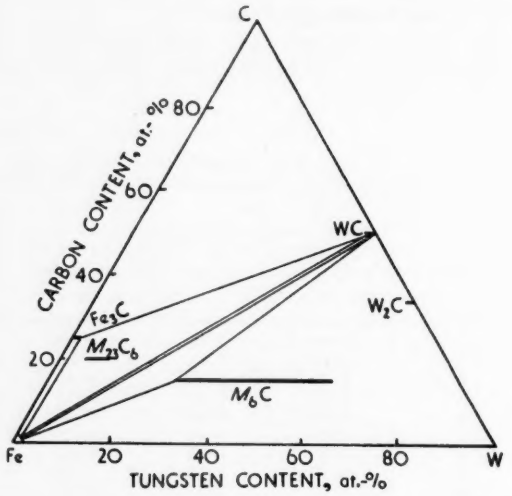


Fig. 4—Stable equilibrium diagram of the system Fe–W–C at 700° C.

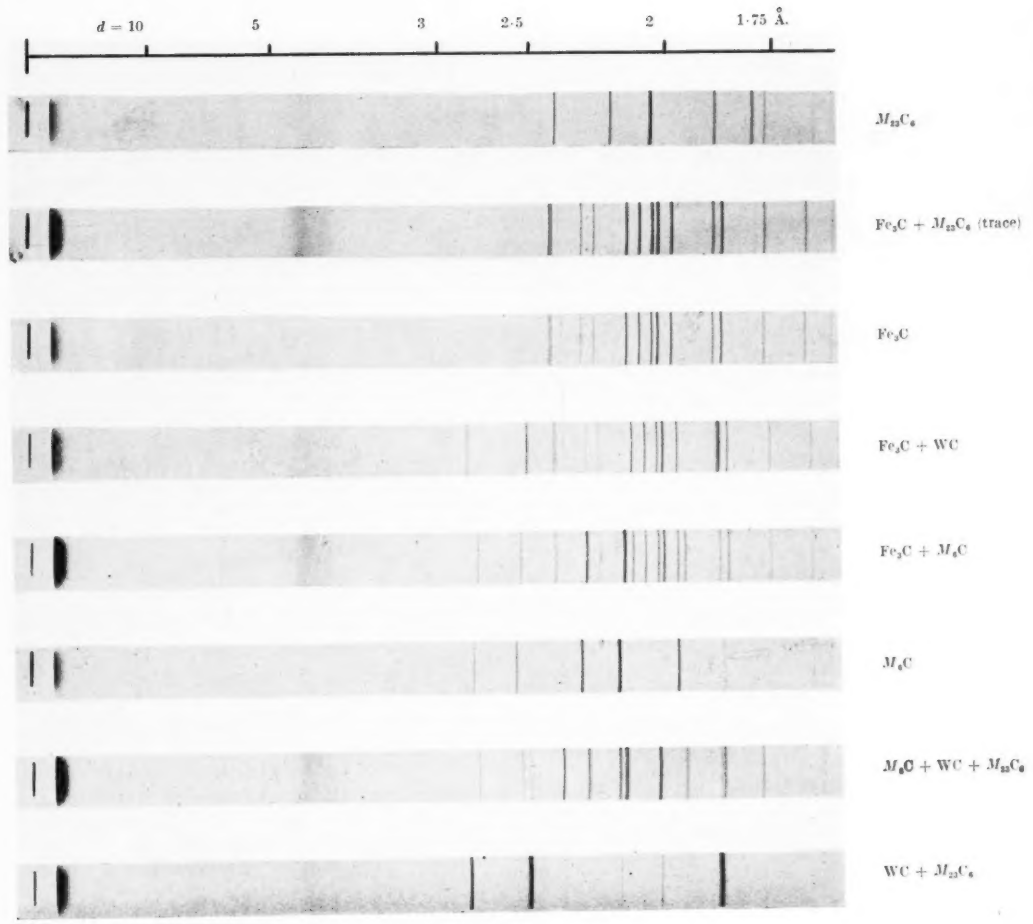


Fig. 5—X-ray photographs of carbides in tungsten steels, CrKα-radiation

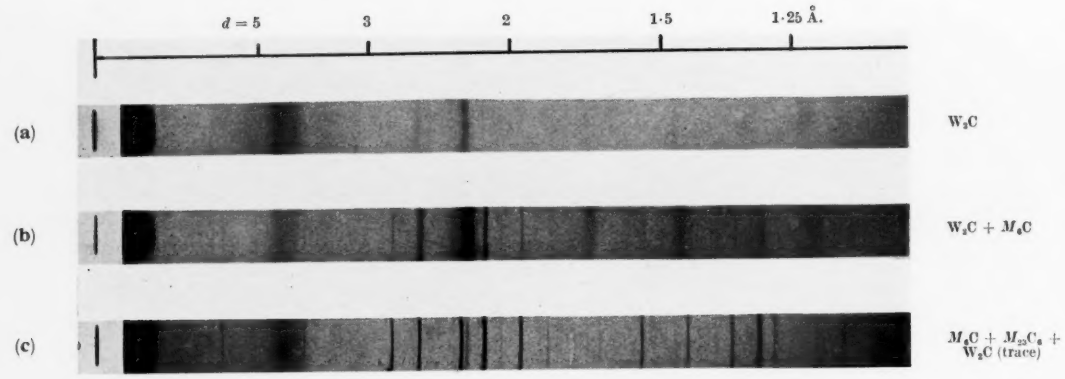


Fig. 6—X-ray photographs of carbides in tungsten steels, CuKα-radiation

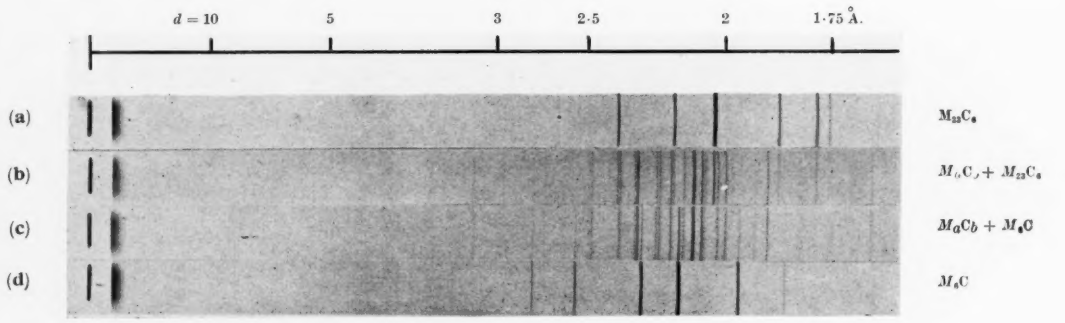


Fig. 7—X-ray photographs of the new iron-molybdenum double carbide M_aC_b, CrKα-radiation

Table V

CARBIDE REACTIONS IN TUNGSTEN STEELS

Carbide Reaction	Steel		X-Ray Analysis of Carbide Residue Extracted from Steel Specimen Quenched from 1300° C. and Tempered at 700° C. for:						
	No.	Atomic Ratio W : C	3 min.	10 min.	1 hr.	5 hr.	25 hr.	500 hr.	2000 hr.
A	*W1*	0·033	Fe_3C	...	...	Fe_3C	$Fe_3C + M_{23}C_6$	...	$Fe_3C + WC$
B	*W2, W3*	0·065–0·15	Fe_3C	...	Fe_3C	$Fe_3C + WC$	$Fe_3C + WC$	...	$Fe_3C + WC$
C	*W4, W5*	0·19–0·23	Fe_3C	...	$Fe_3C + M_6C$	$Fe_3C + M_6C + WC$	$Fe_3C + WC$	...	$Fe_3C + WC$
D	*W6, W7*	0·40–0·67	Fe_3C	...	$Fe_3C + M_6C$	$M_6C + Fe_3C + M_{23}C_6 + WC$	$M_{23}C_6 + M_6C + WC$	$WC + M_{23}C_6$	$WC + Fe_3C$
E	*W8, W9*	1·32–1·57	W_2C*	$W_2C + M_6C$	M_6C	$M_6C + M_{23}C_6$	$M_6C + WC$	$WC + M_6C$	$WC + M_6C$
F	*W10*	1·96	W_2C*	$W_2C + M_6C$	M_6C	M_6C	$M_6C + WC$	$M_6C + WC$	$M_6C + WC$

* Fe_3C appears before W_2C if the steel specimen is tempered at a lower temperature

a niobium-stabilized 18/8 stainless steel without mentioning its composition. The present author* has synthesized W_3Mn_3C ($a = 11·13$Å.), Mo_3Mn_3C (11·10Å.), Nb_3Cr_3C (11·49Å.), and Zr_3V_3C (12·12Å.); the Nb_3Cr_3C is probably the M_6C found by Binder. Many double oxides and nitrides of transition metals also have this structure and composition, as found by Karlsson[26, 27]; *e.g.*, Ti_3Cu_3O, Ti_3Ni_3O, Ti_3Co_3O, Ti_3Fe_3O, Ti_3Mn_3O, Zr_3V_3O, Zr_3Cr_3O, Ti_3Fe_3N, and Ti_3Mn_3N.

CARBIDE REACTIONS IN TUNGSTEN STEELS

Quenched and Tempered Steel Specimens

Steel specimens were quenched from 1300° C. and then tempered at 700° C. for periods ranging from 3 min. to 2000 hr. The carbide phases found after tempering are shown in Table V. The final tempering products are the same as those obtained by long annealing, and the first carbide found in all cases was cementite. Six different carbide reactions were observed:

Reaction A—Carbide $M_{23}C_6$ appears in small amounts and vanishes when WC begins to appear (*see* Figs. 5*b* and *d*).

Reaction B—No intermediate product occurs as equilibrium ($Fe_3C + WC$) is reached.

Reaction C—Carbide M_6C occurs as an intermediate product of tempering; $M_{23}C_6$ never appears, although it occurs after isothermal transformation in the same steel (*see* Fig. 5*e*).

Reaction D—Carbide M_6C occurs first and $M_{23}C_6$ later, as intermediate products of tempering; Fe_3C vanishes after short tempering but later reappears after longer tempering at 700° C. (*see* Figs. 5*e*, *g*, and *h*).

Reaction E—Carbides W_2C and $M_{23}C_6$ occur as intermediate products of tempering (*see* Figs. 6*a*, *b*, and *c*).

Reaction F—As reaction *E*, except that there is no temporary formation of $M_{23}C_6$.

Appearance, Disappearance, and Reappearance of Fe_3C

During tempering, Fe_3C always forms first (possibly after ε-iron carbide[5]). Since Fe_3C dissolves little tungsten and the carbon content of ferrite is extremely low, the atomic ratio W: C of the ferrite, resulting from the precipitation of cementite from martensite, is high. Consequently, the tungsten-containing carbides begin to appear after Fe_3C is formed. Carbide Fe_3C redissolves to supply carbon for the further formation of the tungsten-containing carbides, and if the tungsten content of the steel is high it disappears completely (reactions *D*, *E*, and *F*). Naturally, the rate of disappearance of Fe_3C increases as the atomic ratio W: C in the steel increases.

In steel having an atomic ratio W: C smaller than unity, nearly all the tungsten is bound in WC after equilibrium is reached. The excess carbon then combines with iron, and Fe_3C thus reappears (reaction *D*).

Tungsten-Containing Carbides Appearing Immediately After Fe_3C

The higher the atomic ratio W: C of the ferrite, the higher will be the atomic ratio W: C of the tungsten-containing carbide formed immediately after Fe_3C. As is shown in Table V, the carbides appearing immediately after Fe_3C in steels with increasing atomic ratio W: C are $M_{23}C_6$ ($Fe_{21}W_2C_6$), WC, M_6C (Fe_4W_2C), and W_2C, arranged in order of increasing atomic ratio W: C. This explains why, in reaction *B*, WC forms directly after Fe_3C without any intermediate step.

Both M_6C (Fe_4W_2C) and W_2C have the same atomic ratio W: C, but the former occurs in steels with lower atomic ratio W: C than W_2C does. This is possibly connected with the fact that the latter needs a greater concentration of tungsten atoms; in Fe_4W_2C the atomic percentage of tungsten is 28·6, whereas in W_2C it is 66·7. In steels with medium tungsten content it is easier to form M_6C than to form W_2C, since iron atoms also take part in the formation of this double carbide and not all of them need to diffuse out. Only when the tungsten content of the steel is sufficiently high does W_2C form directly after Fe_3C.

In steel *W*8, the carbide formed after W_2C is M_6C and after M_6C is $M_{23}C_6$, in order of decreasing atomic ratio W: C (reaction *E*). This can be explained in the same way as the isothermal transformation of steel *W*7 at 700° C. (p. 366).

* Unpublished work.

Metastable Carbide $M_{23}C_6$

Carbide $M_{23}C_6$ ($Fe_{21}W_2C_6$) has an atomic ratio W: C of 0·33 (in practice it is much lower, as $Fe_{21}W_2C_6$ is an ideal composition and usually only part of the tungsten sites in this carbide are occupied by tungsten atoms) and therefore, for each tungsten atom, it is more effective in consuming carbon (or Fe_3C) than is M_6C (Fe_4W_2C). Otherwise, Fe_3C would not completely disappear in steels *W*6 and *W*7 (reaction *D*), since there is not enough tungsten to combine all the carbon to form M_6C. Carbide $M_{23}C_6$ occurs temporarily in reaction *E* because the atomic ratio W: C of the steels (*W*8 and *W*9, 1·32 and 1·57) is smaller than that of M_6C (2·0).

Metastable Carbide M_6C

Owing to the increase of the atomic ratio W: C of ferrite caused by the formation of Fe_3C, carbide M_6C can form in steels having an atomic ratio W: C as low as 0·19 (reaction *C*). But it disappears later, probably because tungsten is deficient and WC will absorb the available tungsten.

In steel *W*4, $M_{23}C_6$ is the only carbide present after the completion of the isothermal transformation, and it contains 14·2% of tungsten. However, $M_{23}C_6$ never occurs in the tempered specimens of this steel, possibly because, after the formation of Fe_3C, the ferrite has too high an atomic ratio W: C. All the tungsten is concentrated in a smaller amount of M_6C with higher tungsten content.

Metastable Carbide W_2C

The formation and disappearance of W_2C have already been discussed (p. 369), and the technical importance of this carbide will now be considered.

It was found[28] that the secondary hardening of tungsten hot-working and high-speed steels is associated with the precipitation of W_2C. The addition of chromium and cobalt makes the growth of this carbide very sluggish. When M_6C begins to appear the hardness of the steel is already rather low. The details of this investigation will be published shortly.

Wood and Wainwright's transition compound[16] is probably W_2C.

Rate of Formation of WC

Carbide WC forms, in general, after tempering for 5–24 hr. at 700° C., whilst 50–100 hr. annealing is needed for it to appear after complete isothermal transformation at the same temperature. This is probably due to the internal stresses induced by quenching, since it is generally known that solid transformations can be accelerated by deformation and the application of stresses. Hultgren and Edström[29] have also shown that graphite in malleable cast iron forms more quickly and in larger numbers of nuclei if it is pre-quenched.

Section III—CARBIDES IN MOLYBDENUM STEELS

EQUILIBRIUM DIAGRAM OF THE SYSTEM Fe–Mo–C BELOW A_1

Carbides in the System Fe–Mo–C

All the carbides found in the system Fe–W–C have their analogues in this system. The existence of MoC, however, has been very much disputed. It has not been found by Westgren and Phragmén[30] and Takei[31] in their studies of the binary system Mo–C. The crystal structures reported for MoC by Sykes, van Horn, and Tucker[32] and Hoyt[33] seem to be incorrect. The only evidence of the existence of MoC with a structure isomorphous with WC was Tutiya's finding,[34] a simple hexagonal lattice with one molybdenum atom in the unit cell ($a = 2·901$Å., $c = 2·786$Å.).

Table VI

CARBIDES IN THE SYSTEM Fe–Mo–C

Carbide	Space Lattice	Lattice Dimensions, Å.	No. of Formula-Units per Unit Cell	Remarks
Fe_3C	*See* Table I			
MoC	Hexagonal	$a = 2·898$ $c = 2·809$	1	Found by present author in many molybdenum steels
	Hexagonal	$a = 2·932$ $c = 10·97$	4	First found by Kuo and Hagg[35] when carburizing molybdenum powder with CO between 600° and 775° C. It has not been found in steels
Mo_2C*	Hexagonal	$a = 3·022$ $c = 4·724$	1	Found by Takei[45] in many molybdenum steels
$M_{23}C_6$	*See* Table III			
M_6C	*See* Table III			
M_aC_b	Unknown			Found by present author in this investigation. Lashko and Nesterova[4] reported Fe_nMo_mC in 1951, and this may represent the same carbide

* A cubic modification of this carbide with NaCl-type structure has recently been reported by Lander and Germer[43] and Nowotny and Kieffer[44]

Kuo and Hägg[35] recently found a low-temperature modification of MoC with four molybdenum atoms in one hexagonal cell of the space group $C6/mmc$. They also confirmed Tutiya's report of the existence of MoC with WC structure, which has been found in many molybdenum steels (*see* p. 372).

A third Fe–Mo double carbide has been found in this investigation, but its structure is still unsolved. It seems to have a very limited composition range between $M_{23}C_6$ and M_6C, and attempts to determine its exact composition have failed. It will therefore be denoted by the formula M_aC_b, where a and b are undetermined integrals. During the preparation of this paper, Lashko and Nesterova[4] also reported an Fe–Mo double carbide, which they called Fe_nMo_mC, in a molybdenum steel containing 4·1% Mo and 0·33% C, after tempering for 20 hr. at 600° C. Their Fe_nMo_mC is probably the present author's M_aC_b, as the latter has also been found in this investigation in a steel with similar composition (steel *M*7). This is not the X-phase reported by Andrews.[36]

Lashko and Nesterova[4] also reported a γ′ phase of NaCl-type lattice ($a = 3·60$Å.) in the early stage of tempering of high-molybdenum steels, as they did in the case of high-chromium steels (*see* p. 364). This phase, as discussed earlier, is probably the retained austenite undissolved in the electrolyte.

The crystallographic data of the carbides of this system are shown in Table VI.

Isothermally Transformed Steel Specimens

The compositions of the steels and the X-ray analyses of the carbides obtained by isothermal transformation and tempering are shown in Table VII. The carbides formed during isothermal transformation in these steels are similar to those obtained in tungsten steels, except for the presence of Mo_2C in steels with medium molybdenum contents. Perhaps this is because Mo_2C has a certain stability; this will be discussed in connection with the carbide reactions (p. 372).

Carbide $M_{23}C_6$ may contain from 2% to 8% of molybdenum, and M_6C contains 50–60% of molybdenum. The cube edges of these two carbides are about 10·52Å. and 11·08Å., respectively.

Carbides M_aC_b and MoC have not been identified in steels immediately after the isothermal transformation.

Equilibrium Diagram

After tempering for 5000 hr. at 700° ., equilibrium was reached in all steels except *M*6 (*see* Table VII). Carbides Mo_2C and $M_{23}C_6$ behaved in the same way

Table VII

CARBIDE REACTIONS IN MOLYBDENUM STEELS

Steel No.	Steel C, %	Steel Mo, %	Steel Atomic Ratio Mo:C	X-Ray Analysis of Carbide Residue Extracted from Steel Specimen: After Isothermal Transformation* for 24 hr. at 700° C.	After Tempering* at 700° C. for: 3 min.	1 hr.	25 hr.	100 hr.	500 hr.	2000 hr.	5000 hr.
*M*1	1·04	0·32	0·039	Fe_3C	Fe_3C	Fe_3C	Fe_3C	Fe_3C	Fe_3C	Fe_3C	Fe_3C
*M*2	0·46	0·32	0·087	$M_{23}C_6$	Fe_3C	Fe_3C	Fe_3C	$Fe_3C + M_{23}C_6$	$Fe_3C + M_{23}C_6$	$Fe_3C + MoC$	$Fe_3C + MoC$
*M*3	0·54	0·54	0·125	$M_{23}C_6$	Fe_3C	Fe_3C	Fe_3C	$Fe_3C + MoC$	$Fe_3C + MoC$	$Fe_3C + MoC$	$Fe_3C + MoC$
*M*4	0·54	0·82	0·190	$M_{23}C_6$	Fe_3C	Fe_3C	$Fe_3C + Mo_2C$	$Fe_3C + Mo_2C$	$Fe_3C + MoC$	$Fe_3C + MoC$	$Fe_3C + MoC$
*M*5	0·23	0·80	0·435	$M_{23}C_6 + Mo_2C$	Fe_3C	$Fe_3C + Mo_2C$	$Fe_3C + Mo_2C + M_aC_b$	$Fe_3C + Mo_2C + M_aC_b + MoC$	$Fe_3C + Mo_2C + M_aC_b + MoC$	$MoC + Fe_3C$	$MoC + Fe_3C$
*M*6	0·42	2·85	0·850	$M_{23}C_6 + Mo_2C$	$Fe_3C + Mo_2C$	$Fe_3C + Mo_2C + M_aC_b$	$M_aC_b + Mo_2C + Fe_3C$	$M_aC_b + Mo_2C + M_{23}C_6$	$M_aC_b + M_{23}C_6 + MoC$	$M_aC_b + M_{23}C_6 + MoC$	$M_aC_b + M_{23}C_6 + MoC$
*M*7	0·42	4·08	1·22	$Mo_2C + M_6C + M_{23}C_6$	Mo_2C	Mo_2C	$Mo_2C + M_aC_b$	$M_aC_b + Mo_2C$	$M_aC_b + Mo_2C$	$M_aC_b + M_6C + MoC$	$MoC + M_6C$
*M*8	0·38	5·15	1·70	$M_6C + Mo_2C$	Mo_2C	Mo_2C	$Mo_2C + M_6C$	$M_6C + M_aC_b + Mo_2C$	$M_6C + M_aC_b$	$M_aC_b + M_6C + MoC$	$M_6C + MoC$
*M*9	0·18	3·88	1·72	M_6C	Mo_2C	$Mo_2C + M_6C$	$M_6C + Mo_2C$	M_6C	M_6C	$M_6C + MoC$	$M_6C + MoC$
*M*10†	0·23	6·10	3·32	M_6C	$Mo_2C + M_6C$	$M_6C + Mo_2C$	M_6C	M_6C	M_6C	M_6C	M_6C
*M*11†	0·20	5·09			...	...	...	M_6C	M_6C	$M_6C (+ Fe,Mo_3)$	$M_6C (+ Fe,Mo_3)$

* Austenitizing temperature: 1300° C. for steels *M*1 to *M*8; 1200° C. for steels *M*9 to *M*11

† Carbides were not completely dissolved after austenitizing at 1200° C. At higher temperatures ferrite began to appear

as their isomorphs in tungsten steels, disappearing after long tempering. As can be seen in steels *M*5, *M*7, and *M*8, M_aC_b is also a metastable carbide and occurs only as an intermediate product during tempering and annealing.

Carbide MoC (with WC structure) occurred in all steels having an atomic ratio Mo: C between 0·087 and 1·72. In this respect this carbide resembles WC, although it has not previously been reported in molybdenum steels. Carbide MoC has not been found in steel *M*1, perhaps because of the small amount present in the carbide mixture. However, the possibility that some molybdenum may dissolve in Fe_3C without forming its own carbide should not be overlooked. In fact, Kuo and Hultgren[21] have reported a partition ratio of molybdenum between cementite and ferrite of 8 : 1 in low-molybdenum steels in equilibrium at 700° C. The lattice dimensions of MoC extracted from molybdenum steels are $a = 2·892$Å., $c = 2·802$Å.—about the same as those of pure MoC.

The carbides present in equilibrium with ferrite, and their relations with the atomic ratio Mo: C of the steel, are analogous to those in tungsten steels. Carbide M_6C exists alone if the ratio is larger than 3·3. The lower limit of this ratio is not determined, but is possibly between 2 and 3 (*cf.* Fe_4Mo_2C and Fe_3Mo_3C). This confirms the statement made on p. 368 that M_6C is a stable carbide if the atomic ratio W: C is larger than unity, and that it will occur alone if the ratio is larger than 3. The intermediate phase Fe_7Mo_6 begins to appear when the atomic ratio Mo: C reaches 5 (steel *M*11). The equilibrium diagram in this system is similar to that of the system Fe–W–C.

CARBIDE REACTIONS IN MOLYBDENUM STEELS

Quenched and Tempered Steel Specimens

The carbide reactions in quenched and tempered specimens (*see* Table VII), except in two special cases, are similar to the carbide reactions found in tungsten steels.

Metastable Carbide Mo_2C

In discussing the carbides obtained by isothermal transformation (p. 371), it was stated that Mo_2C has a certain stability, at least in comparison with W_2C, which has not been found in steels after isothermal transformation. Carbide Mo_2C was known to occur in steels with a molybdenum content as low as 0·4%.[37] It disappears at a very low rate; comparing steels *W*9 and *M*9 (both having an atomic ratio of metal to carbon of about 1·6), at 700° C.carbide W_2C vanishes after 1 hr., whereas Mo_2C vanishes after 25 hr. and in low-molybdenum steels disappears only after tempering for 2000 hr. at this temperature.

This perhaps accounts for the phenomenon that in steel *M*4 the carbide sequence is Fe_3C, $Fe_3C + Mo_2C$, and Fe_3C + MoC, compared with the sequence Fe_3C, $Fe_3C + M_6C$, and Fe_3C + WC which is found in tungsten steels (*see* reaction *C*).

Kuo and Hägg* have found that, in carburizing molybdenum powder with carbon monoxide, once Mo_2C is formed it never disappears during prolonged carburization, and MoC can only be produced by preventing the formation of Mo_2C in the beginning of carburization. In contrast to Mo_2C, W_2C has never been found in carburizing tungsten powder. This also serves as evidence of the high stability of Mo_2C.

Like W_2C in tungsten high-speed steels, the formation of Mo_2C accounts for the secondary hardening of molybdenum high-speed steels.[28]

Metastable Carbide M_aC_b

By studying carbide reactions in steels *M*6, *M*7, and *M*8, it can be concluded that M_aC_b has a composition somewhere between those of $M_{23}C_6$ and M_6C. After Mo_2C, M_aC_b occurred in a steel (*M*7) with a lower atomic ratio Mo: C, whilst M_6C occurred in a steel (*M*8) with a higher atomic ratio. Applying the same reasoning as that used to explain the different carbides appearing after Fe_3C in tungsten steels, it is to be expected that M_aC_b has a lower atomic ratio Mo: C than M_6C has. In steel *M*8, the carbide appearing after Mo_2C was M_6C and after M_6C was M_aC_b. In steel *M*6, the carbide appearing after M_aC_b was $M_{23}C_6$. Hence it is probable that M_aC_b has an atomic ratio Mo: C between those of M_6C and $M_{23}C_6$ (*cf.* reaction *D* in tungsten steels).

Carbide M_aC_b has been produced by carburizing M_6C, and Mo_2C was also produced. Consequently, M_aC_b contains less molybdenum than M_6C does.

Carbide M_aC_b has not been found in molybdenum steels containing 1% of chromium.

The X-ray powder photographs of this carbide mixed with $M_{23}C_6$ or M_6C are shown in Fig. 7. Attempts at indexing its X-ray diffraction pattern have so far not succeeded. Observing the high line density in the neighbourhood of *d* (interplanar spacing) = 2Å., the symmetry of this double carbide cannot be very high. Since this carbide appears only after other carbides have been formed and after a long time at 700° C., it is doubtful whether single crystals of this phase could ever be obtained. It is hoped to index the powder pattern in the future.

Section IV—GENERAL DISCUSSION

CHROMIUM *versus* MOLYBDENUM AND TUNGSTEN

The behaviours of these three elements in steels, chromium on the one hand and molybdenum and tungsten on the other, are quite different, as is evidenced by the types of carbide formed and the carbide reaction sequence.

Molybdenum and tungsten form four carbides in common; *i.e.*, MoC (WC), Mo_2C (W_2C), $M_{23}C_6$, and M_6C. The carbides that are found only in the system Fe–Mo–C are the low-temperature modification of MoC and the new iron–molybdenum double carbide M_aC_b. Analogous carbides may also exist in the system Fe–W–C, but none has yet been detected. On the other hand, chromium forms only one carbide, $Cr_{23}C_6$, in common with molybdenum and tungsten. However, $Cr_{23}C_6$ is a binary phase whilst $M_{23}C_6$ is a ternary one. The solubility of these elements in Fe_3C

* Unpublished work.

also differs. Chromium can replace iron in Fe_3C up to 18%, whereas the solubility of molybdenum or tungsten in Fe_3C is very limited.

In tempering chromium steels, the concentration of chromium in Fe_3C takes place first, and the transformation of Fe_3C into Cr_7C_3 will not begin until the chromium content of Fe_3C reaches a critical value of about 20%. But by then the chromium content of the ferrite matrix is already very much impoverished. If Cr_7C_3 were formed from the matrix, it should occur more easily in a matrix with a higher chromium content; in other words, immediately after the formation of Fe_3C and before any concentration of chromium in Fe_3C takes place. The experimental results seem to favour the hypothesis that Cr_7C_3 is the transformation product of a reaction of Fe_3C with the matrix. The reaction mechanism of $Cr_7C_3 \rightarrow Cr_{23}C_6$ is not known, but it may follow similar stages.

The kinetics of the carbide reactions in molybdenum and tungsten steels are of a different nature from the above-mentioned reaction. The solubility of molybdenum or tungsten in Fe_3C is small; consequently, the molybdenum or tungsten content of the matrix does not change materially after the formation of Fe_3C. On the contrary, the atomic ratio of Mo: C or W: C in the matrix increases considerably, and thus the molybdenum- or tungsten-containing carbide is able to form. The following arguments substantiate this hypothesis:

(i) Carbide Mo_2C or Fe_4W_2C occurs together with Fe_3C in low-molybdenum or tungsten steel after tempering, and $M_{23}C_6$ is the only carbide present in the same steel after isothermal transformation. If the carbide reaction follows the same mechanism as $Fe_3C \rightarrow Cr_7C_3$, it is difficult to understand why the transformation product of Fe_3C is not $M_{23}C_6$, instead of Mo_2C or Fe_4W_2C, a carbide with a much higher molybdenum or tungsten content.

(ii) In the same tungsten steel (*W*7), the carbide formed after Fe_3C is M_6C and after M_6C is $M_{23}C_6$. This agrees well with the precipitation hypothesis; a carbide containing less tungsten is formed when the atomic ratio W: C of the matrix is reduced, because of the precipitation of a tungsten-rich carbide, to such an extent that further precipitation of this carbide is rendered impossible. But the atomic ratio W: C is still high enough for the formation of a carbide with low tungsten content. It can, of course, be argued that there exists a certain lattice relationship between Fe_3C and M_6C which favours the direct transformation of the former to the latter, and that this relationship does not exist between Fe_3C and $M_{23}C_6$. The crystallography of the carbides is discussed later; at present the author merely wishes to point out that $M_{23}C_6$ does occur after Fe_3C in steels with a very low molybdenum or tungsten content.

(iii) The formation of Mo_2C or W_2C is accompanied by a very pronounced increase in the hardness of the steel. This is a type of precipitation hardening.[28] Direct carbide → carbide transformation would not have such a large effect.

CHEMISTRY *versus* CRYSTALLOGRAPHY

In explaining the observed phenomena in carbide reactions, only the chemical compositions of the iron phase and of the different carbides are considered. Although the explanations given in sections I–III seem plausible, a short discussion from the crystallographic point of view would probably not be out of place.

Lashko and Nesterova[4] claimed, without any theoretical evidence, that there exists a certain crystallographic relationship between ferrite and Fe_3C, and that this is the reason why Fe_3C forms before Cr_7C_3 or Mo_2C. The crystallography of the formation of ε-iron carbide and cementite during the decomposition of martensite has recently been determined by Jack.[5] Carbide Fe_3C is formed at low temperature, and at this temperature only carbon atoms can migrate. A simple count of the number of iron atoms and molybdenum or tungsten atoms in the matrix will make it clear why Fe_3C forms instead of any other carbide.

In discussing the reason why $M_{23}C_6$ forms after isothermal transformation whereas Fe_3C forms after tempering the same steel, Goldschmidt (*loc. cit.*,[2] p. 356) called attention to the fact that $M_{23}C_6$ is related to both austenite and martensite in the respect that the cubo-octahedra which form an integral unit of repeat in this carbide are the simplest possible compromise between a face-centred cube and a body-centred cube. He therefore considered that $M_{23}C_6$ is "formed as a stable alternative to martensite during the decomposition of austenite," and that, similar to martensite, it is "able to form from austenite by a spontaneous transformation of the lattice, not involving atomic diffusion and therefore proceeding instantaneously." This explanation contradicts the present results that when the same steel, with $M_{23}C_6$ as the austenite transformation product at high temperature, is isothermally transformed at a lower temperature, Fe_3C is formed instead of $M_{23}C_6$. It is difficult to see how $M_{23}C_6$ can form without diffusion if the matrix, whether it is austenite or ferrite, does not have the same composition as $M_{23}C_6$ has.

The complex structure of Cr_7C_3, $M_{23}C_6$, or M_6C has commonly been used to explain the retardation effect of chromium, molybdenum, or tungsten on austenite transformation. As shown in this investigation, $M_{23}C_6$ and M_6C, with 116 and 112 atoms respectively in one unit cell, are formed before WC (or MoC), which has only two atoms in one unit cell with a simple hexagonal lattice. Crystal structure does not seem to play an important rôle in determining the rates of formation of the different carbides. If the composition of the mother phase is suitable, the distances the atoms need to move to form the precipitating phase are only of the magnitude of a few Ångström units, and this is not difficult to attain. The present author is of the opinion that it is the chemical composition of the iron phase and the carbide, rather than the crystal structure of the carbide, which is the determining factor in the formation of carbide.

CONCLUSIONS

(1) In the system Fe–Cr–C, the carbides present in equilibrium at 700° C. are Fe_3C, Cr_7C_3, or $Cr_{23}C_6$, in order of increasing chromium content of the alloy. Carbide Cr_3C_2 does not exist in chromium steels.

(2) The stable carbides in the Fe–W–C system at 700° C. are Fe_3C and WC, if the atomic ratio W: C in the steel is smaller than unity; otherwise, M_6C and WC coexist as stable carbides. Carbides $M_{23}C_6$

and W_2C are metastable. When carbides are formed directly from austenite at 700° C. by isothermal transformation, they appear in the order Fe_3C, $M_{23}C_6$, and M_6C as the atomic ratio W: C in the steel increases.

(3) Stable and metastable carbides analogous to those described above are also found in the Fe–Mo–C system. In addition, a new iron–molybdenum double carbide M_aC_b, with unknown structure, has been found in steels containing 1–5% of molybdenum; its composition is probably between those of $Fe_{21}Mo_2C_6$ and Fe_4Mo_2C. This carbide occurs only after long annealing at 700° C. and is metastable. Carbide Mo_2C also occurs in isothermally transformed molybdenum steels of very low molybdenum content (0·4%).

(4) On tempering, Fe_3C with inherited metal composition from the matrix forms first (after Jack's ε-iron carbide) in all steels. On prolonged tempering, however, other alloy carbides are formed at the expense of Fe_3C. Carbide Fe_3C can dissolve as much as 18% of chromium but less than 1–2% of molybdenum or tungsten.

(5) The general carbide reaction in chromium steels is $Fe_3C \rightarrow Cr_7C_3 \rightarrow Cr_{23}C_6$. The carbide reaction $Fe_3C \rightarrow Cr_7C_3$ is found to take place in two stages: the concentration of chromium in Fe_3C to its maximum solubility, and the transformation from Fe_3C into Cr_7C_3. The carbide reaction $Cr_7C_3 \rightarrow Cr_{23}C_6$ may follow the same stages. At high temperatures, $Cr_{23}C_6$ is found to be less stable than Cr_7C_3, the former dissolving first and completely in austenite.

(6) In tungsten steels, the tungsten-containing carbides formed after Fe_3C appear in the order of increasing atomic ratio W:C; *i.e.*, $M_{23}C_6$ ($Fe_{21}W_2C_6$), WC, M_6C (Fe_4W_2C), and W_2C. On prolonged tempering, the carbide formed after W_2C is M_6C and after M_6C is $M_{23}C_6$. The metastable carbides W_2C and $M_{23}C_6$ later disappear completely. Carbide Fe_3C first redissolves in ferrite to supply carbon for the further formation of tungsten-containing carbides, and, if the W: C ratio is smaller than unity, reappears later when all the available tungsten is bound in WC.

(7) The carbide reactions in molybdenum steels are similar to those found in tungsten steels, with the following two exceptions: (*a*) On tempering, the carbides appearing after Fe_3C in steels with increasing atomic ratio Mo: C are $M_{23}C_6$, MoC, and Mo_2C; M_6C appears always after Mo_2C. (*b*) The carbide appearing after M_6C is M_aC_b, not $M_{23}C_6$.

(8) In explaining the observed carbide reactions, the chemical compositions of the carbides and matrix are considered, instead of the crystal structure of the carbides. It is tentatively suggested that Cr_7C_3 is the reaction product of Fe_3C and the matrix, whereas the molybdenum- and tungsten-containing carbides are formed directly from the matrix.

(9) The rate of carbide reaction is found to be higher in the quenched and tempered steel specimens than in the isothermally transformed and annealed specimens. This is probably due to the internal stress induced by quenching. The reaction rate is high in chromium steels and is extremely low in molybdenum steels.

Acknowledgments

The author wishes to express his sincere gratitude to Professor Emeritus A. Hultgren, of the Royal Institute of Technology, Stockholm, and Professor G. Hägg, head of the Institute of Chemistry, Uppsala University, from whom he acquired the knowledge of metallography and X-ray crystallography needed to carry out this investigation. He also thanks them for the laboratory facilities.

Thanks are also due to Dr. M. Tigerschiöld for his interest, and to Mr. H. Nathorst and Mr. K. Hellbom, through whose co-operation many of the steels used in this investigation were kindly supplied by Sandvikens Jernverks AB and Söderfors Bruk.

The author acknowledges gratefully the financial aid from the Swedish Government Technical Research Council and Jernkontoret (The Swedish Ironmasters Association). Permission to publish this paper was granted by the Research Council of Jernkontoret.

APPENDIX

Experimental Details

The steels used in this investigation were made in high-frequency furnaces, and in no case did they contain more than one of the alloy elements chromium, molybdenum, and tungsten. To eliminate the complication introduced by undissolved carbides, the steel specimens were austenitized in a purified nitrogen atmosphere at 1300° C. for 10 min. In special cases, *e.g.*, high-chromium and high-molybdenum steels, lower austenitizing temperatures were used, to avoid the formation of ferrite. After this treatment the steel specimens were either quenched into salt water or transformed isothermally at the desired temperature in a lead bath. Short-time tempering was also carried out in a lead bath. The lead bath was deoxidized with a lead–calcium alloy, containing 1% of calcium, every other day, and the surface of the lead was also well protected by a 10-cm. thick layer of charcoal. Decarburization was negligible, but a thickness of 1 mm. was always ground off each side of the surface of the steel specimens after the heat-treatment, to guard against any decarburization. This is considered of great importance, because the carbides present depend not only on the alloy content but also on the carbon content of the steel. For prolonged heat-treatment the specimens were always sealed in evacuated silica tubes, to avoid any decarburization that may happen in the lead bath. After the heat-treatment the specimens were cooled rapidly in salt water.

The method for the electrolytic extraction of carbide has been discussed elsewhere.[21] It should be pointed out, however, that in tungsten steels all the tungsten dissolved in the iron matrix is recovered in the anode residue as oxides, since tungsten is not soluble in hydrochloric acid. The anode residue from tungsten steels was therefore washed three times with hot 10% sodium hydroxide.

The carbide phase-analyses were made from powder

photographs taken with Cu Kα-radiation in a Guinier-type camera, using a bent quartz monochromator. In special cases, however, Cr Kα-radiation was also used. It was estimated that the X-ray powder method used in this investigation is capable of detecting 1% of WC, W_2C, MoC, Mo_2C, or M_6C, and 2–3% of Fe_3C, $M_{23}C_6$, or Cr_7C_3, in the carbide mixtures. However, this high accuracy cannot always be obtained, owing to the exceedingly small particle size of the carbide when it is freshly formed. The fact that some carbides may be lost during electrolytic extraction should also be considered.

References

1. D. R. Banyard: Report No. MG/A/8/46: 1947, London, British Iron and Steel Research Association.
2. H. J. Goldschmidt: *J. Iron Steel Inst.*, 1948, vol. 160, pp. 345–362.
3. H. J. Goldschmidt: *Nature*, 1948, vol. 162, p. 855.
4. N. F. Lashko and M. D. Nesterova: *Izvestiya Akademii Nauk S.S.S.R., Seriya Fizicheskaya*, 1951, vol. 15, pp. 67–71.
5. K. H. Jack: *J. Iron Steel Inst.*, 1951, vol. 169, pp. 26–36.
6. A. Hultgren: *Trans. Amer. Soc. Metals*, 1947, vol. 39, pp. 915–1005.
7. T. Heczko: *Berg- hüttenmänn. Monatsh.*, 1947, vol. 92, pp. 125–139.
8. W. Dannöhl: *Stahl u. Eisen*, 1949, vol. 69, pp. 85–86.
9. A. Westgren, G. Phragmén, and T. Negresco: *J. Iron Steel Inst.*, 1928, No. I, pp. 383–400.
10. W. Tofaute, C. Küttner, and A. Büttinghaus: *Arch. Eisenhüttenwesen*, 1935–36, vol. 9, pp. 607–617.
11. T. Lyman and A. R. Troiano: *Trans. Amer. Soc. Metals*, 1946, vol. 37, pp. 402–448.
12. S. Z. Bokshteĭn: *Zhurnal Tekhnicheskaya Fiziki*, 1949, vol. 19, pp. 532–541.
13. F. Wever and W. Koch: *Arch. Eisenhüttenwesen*, 1950, vol. 21, pp. 143–152.
14. R. W. Balluffi, M. Cohen, and B. L. Averbach: *Trans. Amer. Soc. Metals*, 1951, vol. 43, pp. 497–525.
15. J. R. Lane and N. J. Grant: *Ibid.*, 1952, vol. 44, pp. 113–137.
16. W. A. Wood and C. Wainwright: *Phil. Mag.*, 1932, vol. 14, Series 7, pp. 191–198.
17. H. J. Goldschmidt: *J. Iron Steel Inst.*, 1952, vol. 170, pp. 189–204.
18. H. Krainer: *Arch. Eisenhüttenwesen*, 1950, vol. 21, pp. 39–41.
19. B. M. Rosenbaum: Technical Note No. 1580: July, 1948, National Advisory Committee for Aeronautics.
20. V. Adelsköld, A. Sundelin, and A. Westgren: *Zeitschrift für anorganische und allgemeine Chemie*, 1933, vol. 212, pp. 401–409.
21. A. Hultgren and others: *Jernkontorets Ann.*, 1951, vol. 135, pp. 402–494.
22. V. Berg and H. Krainer: *Proceedings of the First International Powder Metallurgy Conference*, 1948, Graz. *See* also "The Physics of Powder Metallurgy," p. 213: 1951, New York, McGraw-Hill Book Co., Inc.
23. S. Takeda: *Technology Reports of the Tôhoku Imperial University, Sendai*, 1930, vol. 9, pp. 483–514, 627–664; 1931, vol. 10, pp. 42–92.
24. E. J. Sandford and E. M. Trent: Symposium on Powder Metallurgy, *Spec. Rep. No.* 38, pp. 84–91: 1947, London, The Iron and Steel Institute.
25. W. O. Binder: Symposium on the Nature, Occurrence, and Effects of Sigma Phase, *Spec. Tech. Publ. No.* 110, pp. 146–164: 1951, Baltimore, American Society for Testing Materials.
26. N. Karlsson: *Nature*, 1951, vol. 168, p. 558.
27. N. Karlsson: Private communication.
28. K. Kuo: *Research*, 1952, vol. 5, No. 7, pp. 339–340.
29. A. Hultgren and O. Edström: *Jernkontorets Ann.*, 1942, vol. 126, pp. 83–107.
30. A. Westgren and G. Phragmén: *Zeitschrift für anorganische und allgemeine Chemie*, 1926, vol. 156, pp. 27–36.
31. T. Takei: *Science Reports of the Tôhoku Imperial University, Sendai*, 1928, vol. 17, pp. 939–944.
32. W. P. Sykes, K. R. van Horn, and C. M. Tucker: *Trans. Amer. Inst. Min. Met. Eng.*, 1935, vol. 117, pp. 173–189. *Metals Technology*, 1935, vol. 2, Sept., *Tech. Publ. No.* 647.
33. S. L. Hoyt: *Trans. Amer. Inst. Min. Met. Eng.*, 1930, vol. 89, pp. 9–58.
34. H. Tutiya: *Bulletin of the Institute of Physical and Chemical Research, Tokyo*, 1932, vol. 11, pp. 1150–1156.
35. K. Kuo and G. Hägg: *Nature*, 1952, vol. 170, pp. 245–246.
36. K. W. Andrews: *Ibid.*, 1949, vol. 164, p. 1015.
37. W. Crafts and C. M. Offenhauer: *Trans. Amer. Inst. Min. Met. Eng.*, 1943, vol. 154, pp. 361–373. *Metals Technology*, 1943, vol. 10, Feb., *Tech. Publ. No.* 1556.
38. W. A. Wood: *Phil. Mag.*, 1930, vol. 10, Series 71, pp. 659–667.
39. V. I. Arkharov, I. S. Kvater, and S. T. Kiselev: *Bulletin de l'Académie des Sciences de l'URSS, Classe des Sciences Techniques*, 1947, No. 6, pp. 749–756.
40. F. R. Morral, G. Phragmén, and A. Westgren: *Nature*, 1933, vol. 132, pp. 61–62.
41. A. Westgren and A. E. Lindh: *Zeitschrift für physikalische Chemie*, 1921, vol. 98, pp. 181–210.
42. K. Becker: *Z. Metallkunde*, 1928, vol. 20, pp. 427–441.
43. J. J. Lander and L. H. Germer: *Trans. Amer. Inst. Min. Met. Eng.*, 1948, vol. 175, pp. 648–692. *Metals Technology*, 1947, vol. 14, Sept., *Tech. Publ. No.* 2259.
44. H. Nowotny and R. Kieffer: *Zeitschrift für anorganische und allgemeine Chemie*, 1952, vol. 267, pp. 261–264.
45. T. Takei: *Kinzoku no Kenkyu*, 1933, vol. 9. pp. 97–124, 142–173.

Effect of Arsenic and Antimony on Temper-Brittleness

By G. W. Austin, O.B.E., M.A., M.Sc., A. R. Entwisle, M.A., Ph.D., and G. C. Smith, M.A.

SYNOPSIS

The susceptibility of nickel-chrome steels to temper-brittleness is slightly increased by arsenic and greatly increased by antimony. Arsenic in the range 0·2—0·02% has little effect on the tensile properties at room temperature, but the effect of antimony in the range 0·08—0·005% depends upon the precise heat treatment given to the steel.

The tensile properties of the steels in liquid nitrogen can be correlated with their notched-bar impact properties, the correlations being based on Orowan's theory of notch brittleness. Severely embrittled specimens fail in tension with little ductility, the fracture being intergranular (with respect to the prior austenite grains).

739

THE susceptibility of a steel to temper-brittleness can best be assessed by measuring the displacement of transition temperature in a notched-bar impact test. Since this was pointed out by Jolivet and Vidal,[1] the effects of the common alloying elements have been studied. It is now well established that phosphorus increases the susceptibility of alloy steels to temper-brittleness, and that it raises the transition temperature of the steels in the absence of temper-brittleness.[2, 3] Houdremont, Bennek, and Neumeister[4] have shown that arsenic raises the transition temperatures of plain carbon steels, but they did not measure the transition temperatures of the low-alloy steels containing arsenic that they studied. Jolivet and Vidal[1] have shown that antimony has a marked influence on the susceptibility of a 1·5% chromium steel, and the work of Jones and Morgan[5] gave clear evidence that plain carbon steels were much embrittled by additions of antimony.

Both arsenic and antimony are usually present in small amounts in steels. In a list of analyses of eight low-alloy steels used at the National Physical Laboratory,[6] the weight-percentage of arsenic varied between 0·023% and 0·039%, these values being comparable with those for phosphorus. The only values for antimony available were for American steels;[7] the average for 21 steels was 0·01%, with a maximum value of 0·028%, the equivalent values for arsenic being given as 0·01% and 0·034%.

It seemed desirable to investigate the effect of arsenic and antimony on the temper-brittleness of low-alloy steels, since these elements are in the same groups of the periodic table as phosphorus. The effect of bismuth, the other element in the sub-group, which is insoluble in iron,[8] was not investigated.

EXPERIMENTAL PROCEDURE

Two types of steel were used in the investigation, corresponding to British Standards En.23 and En.23 + Mo. Each type was made with four arsenic contents and with two antimony contents. The steels were manufactured by Brown Bayley Steels, Ltd., in a small high-frequency furnace. The four groups were each made from a separate melt. The finished product available for heat treatment and testing was 0·45-in. dia. precision-ground rod. The compositions are given in Table I.

Impact testing was carried out in a standard 120-ft.lb. Izod machine, using the British round specimen (0·45-in. round with a 45° notch 0·130 in. deep). To prevent the interchange of heat between the specimen and the grips of the testing machine, a special pair of grips, lined with Tufnol, was used where necessary. These grips gave reliable results, so long as the specimen was fractured within a short time of being clamped in the machine.

Heat treatment was carried out in such a way as to prevent appreciable attack of the specimen blanks. These were austenitized in a vacuum furnace with a suitable arrangement for oil quenching. Tempering was carried out in neutral salt baths. After treatment, the bars were free from appreciable surface attack, although some discoloration had taken place. After being notched they were ready for testing.

Impact tests were carried out over a range of temperatures to determine the transition temperature. In most cases, 20 specimens, all taken from the same heat-treated batch, were used for each curve. Low temperatures were obtained by using a bath of liquid air or a dry-ice–acetone mixture. Specimens were placed in a close-fitting aluminium block, and cooled down to a temperature below that desired for the test. The block was then suspended in a Dewar flask, and the temperature of the assembly (measured by a thermocouple located at the notch in the specimen) rose slowly with time. When the required temperature

Paper MG/A/85/52 of the Alloy Steels Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association, received 20th November, 1952. The views expressed are the authors', and are not necessarily endorsed by the Committee as a body.

Mr. Austin is Goldsmith's Professor of Metallurgy in the University of Cambridge, Dr. Entwisle is Research Fellow in the Department of Metallurgy, University of Sheffield (formerly of the Department of Metallurgy, University of Cambridge), and Mr. Smith is Demonstrator in Metallurgy in the University of Cambridge.

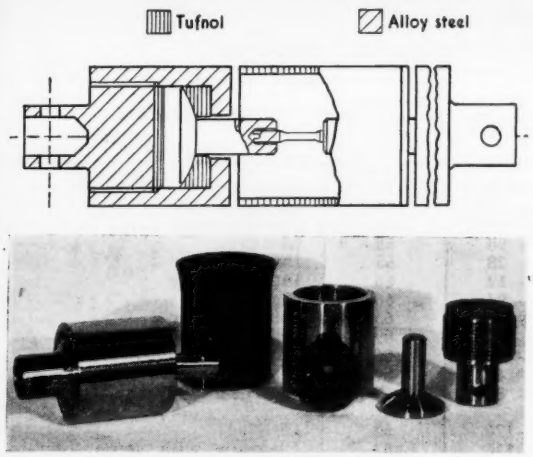


Fig. 1—Grips for the tensile testing of specimens in liquid nitrogen

was reached, the specimen was withdrawn, and the tests were made as quickly as possible. Temperature measurements made with a thermocouple inserted lengthwise down a series of dummy specimens showed that fracture took place not more than 2½° C. from the temperature desired in all cases. Energy-absorption/testing-temperature curves were determined, and in each the transition temperature was taken as that temperature at which the energy absorbed had fallen by 25%. The criterion corresponds roughly with the first appearance of brittle facets in the fracture.

Tensile tests were carried out at room temperature, using a standard tensometer, with a 2-ton beam. The test pieces (cross-sectional area $\frac{1}{40}$ sq. in.) were turned from the larger parts of the fractured impact specimens. Low-temperature tests were carried out on the same machine mounted vertically, with a powered drive. Special thermally insulated grips, shown in Fig. 1, were used; the separate parts are shown in the photograph. The test piece used was $\frac{1}{80}$ sq. in. in cross-section (0·1264 in. dia.) with

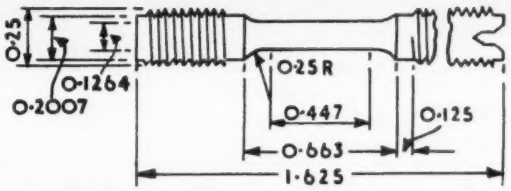


Fig. 2—Tensile specimen for low temperatures (cross-sectional area = $\frac{1}{80}$ sq. in.). All dimensions in inches

threaded ends; its profile is shown in Fig. 2. After machining, the gauge lengths and fillets were polished down on emery paper to 0000. During this process, the gauge length was slightly tapered to the centre, so that the minimum cross-section was about 0·0003 in. less in diameter than the average diameter. This was done to correct any tendency to fracture at the fillets.

The sequence of operations adopted in testing was as follows:

> The test piece was fitted in the grips, and a small load was applied. The system was precooled with liquid oxygen until only slight boiling of the liquid took place (about 15 min.). The oxygen was then removed and replaced by liquid nitrogen. The level of nitrogen in the cup was maintained for 10–15 min., by which time the boiling had practically ceased, and the test was then carried out. The strain rate used was 4 in./in./hr.

Temperature measurements on dummy specimens, with a thermocouple introduced through a hole bored down the centre of one grip, showed that the temperature near the shoulders of the specimen was — 194° C. This means that the maximum temperature variation along the specimen gauge length was 2° C., and was probably less.

EFFECT OF ARSENIC ON IMPACT AND TENSILE PROPERTIES

Specimen blanks of the En.23 steels *S*, *T*, *U*, and *V* were oil-quenched after 1 hr. at 900° C. They were then tempered at 620° C. for 2 hr. Some of each

Table I
COMPOSITIONS OF STEELS USED

Steel No.	Cast No.	British Standards En. No.	C, %	Si, %	Mn, %	S, %	P, %	Cr, %	Ni, %	Mo, %	As, %	Sb, %
S	2245/1	En.23	0·38	0·20	0·43	0·029	0·024	0·66	3·01	0·05	0·021	0·005
T	2245/2		0·41	0·20	0·43	0·029	0·024	0·66	3·02	0·05	0·085	...
U	2245/3		0·42	0·20	0·43	0·029	0·024	0·66	3·00	0·05	0·145	...
V	2245/4		0·40	0·20	0·43	0·029	0·024	0·66	3·01	0·05	0·255	...
W	2246/1	En.23 + Mo	0·42	0·18	0·46	0·029	0·023	0·66	3·02	0·59	0·019	...
X	2246/2		0·41	0·18	0·46	0·029	0·023	0·66	3·03	0·59	0·087	...
Y	2246/3		0·41	0·18	0·46	0·029	0·023	0·66	3·02	0·59	0·15	...
Z	2246/4		0·40	0·18	0·46	0·029	0·023	0·66	3·02	0·59	0·23	...
A	2892/1	En.23	0·42	0·16	0·27	0·028	0·032	0·81	3·04	0·05	...	0·005
B	2892/2		0·40	0·18	0·26	0·025	0·032	0·79	3·04	0·03	...	0·085
C	2892/3	En.23 + Mo	0·41	0·17	0·23	0·029	0·030	0·78	3·05	0·50	...	0·007
D	2892/4		0·41	0·17	0·23	0·038	0·030	0·78	3·05	0·50	...	0·088

Table II

IMPACT PROPERTIES OF STEELS CONTAINING ARSENIC

Designation (En.23)	Hardness, V.D.H.		A.S.T.M. Grain Size	Transition Temperature, °C.	Maximum Impact Value, ft. lb.	Type of Fracture
	O.Q.	As Tested				
S	609	284	8–9	−112	65	Cleavage
SB	609	281	8–9	− 42	65	Intergranular
*SB*8	609	280	8–9	− 25	68	Intergranular
T	610	282	8	− 98	65	Cleavage
TB	610	280	8	− 28	63	Intergranular
*TB*8	610	284	8	− 12	66	Intergranular
U	604	288	8	− 94	66	Cleavage
UB	604	284	8	+ 4	65	Intergranular
*UB*8	604	284	8	+ 12	69	Intergranular
V	596	291	7–8	− 78	66	Mostly cleavage
VB	596	289	7–8	+ 46	64	Intergranular
*VB*8	596	284	7–8	+ 50	69	Intergranular
W (+ Mo)	627	317	8–9	−126	53	Cleavage
WB „	627	313	8–9	−116	53	Cleavage with some intergranular
*WB*8 „	627	306	8–9	−106	58	Cleavage and intergranular
X „	635	316	8	−124	53	Cleavage
XB „	635	318	8	−102	53	Cleavage and intergranular
*XB*8 „	635	301	8	− 98	59	Cleavage and intergranular
Y „	628	316	8	−118	54	Cleavage
YB „	628	313	8	− 96	54	Mostly intergranular
*YB*8	628	305	8	− 90	56	Mostly intergranular
Z „	621	318	8	−114	51	Cleavage
ZB „	621	316	8	− 70	50	Mostly intergranular
*ZB*8 „	621	303	8	− 76	56	Mostly intergranular

batch were embrittled at 525° C. for 4 hr., and some for 8 hr. After notching, the transition curves were determined. The transition temperatures, maximum impact values, grain sizes, diamond pyramid hardnesses, and fracture characteristics of the brittle specimens are given in Table II.

This table shows that as the amount of arsenic increases, the transition temperatures increase in both the embrittled and unembrittled conditions. The change in transition temperature due to the embrittling treatment increases from 70° to 128° C. as the arsenic increases from 0·021% to 0·255%. At the same time, the unembrittled transition temperature is raised from − 112° to − 78° C. A study of the fractures of the brittle specimens showed that the unembrittled steels gave cleavage fractures, but the embrittled steels had intergranular fractures. It can be deduced from this that arsenic impairs the cleavage properties of the steels, as well as increasing the susceptibility to temper-brittleness.

The tensile properties of these steels are given in Table III. As would be expected from the figures

Table III

TENSILE PROPERTIES OF STEELS CONTAINING ARSENIC

Designation (En.23)	Yield Stress	U.T.S.	Fracture Strength	True Fracture Strength	Elongation, %	Red. in Area, %	Type of Fracture
	Tons/sq. in.						
S	48·4	55·6	36·8	105	24	65	Some radial cracks
SB	46·6	54·4	37·6	105	21·7	64	Severe radial cracking
T	47·2	54·4	37·6	109	25	65·5	As for *S*
TB	45·2	55·6	38·0	110	26·5	65·5	„ „ *SB*
U	49·0	57·0	38·4	109·5	25·5	65	„ „ *S*
UB	48·8	56·8	38·4	109	24	65	„ „ *SB*
V	48·8	57·6	39·6	113	25·5	65	„ „ *S*
VB	48·0	56·8	38·4	109	26	65·5	„ „ *SB*
W	54·4	61·2	44·8	112	24	62	Cup and cone
WB	54·2	61·6	44·4	117	...	62	Slight radial cracking
X	54·4	62·4	43·2	117	25	63	„ „ „
XB	56·8	64·0	46·4	116	24	61	Marked radial cracking
Y	54·4	62·0	42·4	115	23	63	„ „ „
YB	54·4	62·0	45·6	114	24·5	60	
Z	55·2	62·4	46·4	116	24	62	„ „ „
ZB	54·4	62·4	45·6	114	24	62	Severe radial cracking

Table IV

IMPACT PROPERTIES OF STEELS CONTAINING ANTIMONY

Designation	Hardness, V.D.H.		A.S.T.M. Grain Size	Transition Temperature, °C.		Maximum Impact Value, ft. lb.	Type of Fracture
	O.Q.	As Tested		T	ΔT		
En.23 *A*	625	303	8	−100	62	52	Cleavage
En.23 *AB*	625	301	8	− 38	62	51	Intergranular
En.23 *B*	623	306	7	− 24	>325	42	Mostly intergranular
En.23 *BB*	623	305	7	>+300	>325	?	Intergranular
En.23 + Mo *C*	628	330	7–8	−115	5	44	Cleavage
En.23 + Mo *CB*	628	331	7–8	−110	5	43	Mostly cleavage
En.23 + Mo *D*	622	327	7	− 96	216	44	Cleavage
En.23 + Mo *DB*	622	327	7	+120	216	27	Intergranular

for the maximum impact value, there is no large trend in the results, either due to the arsenic or to the embrittling. The fracture stress (load at fracture/final area) tends to be slightly larger in the steels containing most arsenic. The appearance of the fractures was found to vary with the heat treatment given to the steels. The embrittled steels showed a more pronounced tendency to star-type failure than those not embrittled.

The En.23 + Mo steels (*W–Z*) were oil-quenched in batches after 1 hr. at 900° C. They were tempered at 650° C. for 2 hr., and some specimens from each batch were embrittled at 525° C. for 4 hr. The transition temperatures, etc., were determined and the results are given in Table II. Here the effect of the arsenic is very marked, the embrittlement increasing rapidly with the amount of arsenic present. It is also clear that arsenic raises the transition temperature, even in the absence of temper-brittleness, as can be seen from the descriptions of the fractures. It has already been shown[9] that the appearance of the brittle fracture can be used as a sensitive criterion for temper-embrittlement. Since all the unembrittled brittle fractures were cleavage, it seems safe to conclude that arsenic impairs the impact properties, irrespective of temper-brittleness. Comparison with the steels free from molybdenum indicates that molybdenum lowers the transition temperatures of the unembrittled steels, despite the differences in hardness.

The tensile properties of these steels were determined (*see* Table III), and show little change due to the composition or the heat treatment.

The appearances of the fractures of the tensile specimens showed interesting features. Typical star-type fracture, in which a series of radial cracks were superimposed on the fracture, occurred in most cases. The more pronounced these cracks were, the less was the width of the cup-and-cone part of the fracture. Hollomon[10] concluded that this type of fracture was associated with temper-embrittlement (in a Mn–Mo steel). With the steels considered in this paper, the radial cracks became more prominent as the amount of arsenic increased and as the embrittling treatment was lengthened. Molybdenum seemed to decrease the tendency towards this type of fracture.

From these two series of tests, it was concluded that:

(i) Arsenic increases the transition temperature of tempered martensite

(ii) Arsenic increases the temper-brittleness of tempered martensite

(iii) The effectiveness of molybdenum in suppressing temper-brittleness depends on the arsenic content of the steel.

These conclusions are very similar to those reached by Baeyertz, Craig, and Sheehan[2, 3] with respect to the effect of phosphorus on low-chromium and low-manganese steels. The effect of arsenic is much less than that of phosphorus in all three respects. The results also show that arsenic has no appreciable effect on the room-temperature tensile properties of the steels.

EFFECT OF ANTIMONY ON IMPACT AND TENSILE PROPERTIES

Steels *A*, *B*, *C*, and *D*, containing antimony, were given heat treatments identical with those for the corresponding arsenical steels, *i.e.*, O.Q. after 1 hr. at 900° C., tempered at 620° C. (650° C. for the steels containing molybdenum) for 2 hr., and half of each batch embrittled 4 hr. at 525° C. The transition temperatures and other relevant data are given in Table IV. The displacement ΔT for steel *A* corresponds closely with that for steel *S* (*see* Table II). The addition of antimony raises the unembrittled transition temperature by 76° C., and increases the displacement due to the embrittling treatment from 62° to > 325° C. The transition curve for the embrittled antimony steels could not be determined because it was so high. It is immediately obvious, however, that antimony exerts an overwhelming influence on temper-brittleness in the En.23 steel. For the steel containing molybdenum, the same large effect is shown; it is so large that the beneficial effect of the molybdenum is more than counteracted by the antimony (*cf.* *D* and *DB* with *A* and *AB*). It is also evident that the antimony raises the unembrittled transition temperatures. For the steel not containing molybdenum, some of this effect must be attributed to embrittlement taking place during the initial tempering treatment, as the brittle fracture of steel *B* was found to be mostly intergranular. For steels *C* and *D*, however, the antimony exerts its influence on the cleavage strength.

The maximum impact value of steel *D* falls considerably on embrittlement. This is reflected in the

Table V

TENSILE PROPERTIES OF STEELS CONTAINING ANTIMONY

Designation		Yield Stress	U.T.S.	Fracture Strength	True Fracture Strength	Elongation, %	Red. in Area, %	Type of Fracture
		Tons/sq. in.						
En.23	*A*	50·4	60·0	43·2	113·7	24	62	Cup and cone with slight radial cracking
	AB	51·4	60·3	43·1	115	24	62·5	Slight cup and cone with well developed radial cracks
	B	51·7	60·3	43·1	105·3	22	52	Cup and cone with slight radial cracking
	BB	52·3	60·7	59·9	72·3	10	17·2	Intergranular flat transverse fracture
En.23 +Mo	*C*	56·9	65·0	47·3	124·5	22	62	Cup and cone
	CB	58·4	67·2	48·0	123·1	22	61	Cup and cone with very slight radial cracking
	D	57·6	65·7	50·0	116·3	20	57	Cup and cone
	DB	59·2	66·8	60	103·5	18	42	Cup and cone, with very slight radial cracking

tensile properties, given in Table V, which were determined in the same way as for the steels containing arsenic. These results show that the effect of antimony is marked, both in the embrittled and unembrittled conditions. The yield stress and ultimate tensile strength are hardly affected, but the elongation and reduction in area are lowered, and the true stress at fracture is reduced. The En.23 steel containing antimony showed only 17% reduction in area when embrittled, and gave a flat intergranular fracture very soon after the maximum load was reached. This implies that when temper-embrittlement is sufficiently severe, temper-brittleness can be detected in tensile tests.

Since the difference in response to the embrittling treatment between the steels containing antimony and those free from this element was so great, it seemed necessary to make sure that the phenomenon was indeed temper-brittleness. The most convenient method was to determine whether the reaction was reversible as far as the impact properties were concerned. A series of specimens of steel *B* (En.23 + antimony) were oil-quenched from 900° C., tempered for 2 hr. at 620° C., and embrittled for 4 hr. at 525° C. They were then reheated to 620° C. for varying lengths of time, and the room-temperature impact values were determined; these are given in Table VI.

Although a complete reversal of embrittlement was not obtained, the impact value was very much improved by the treatment at 620° C. The fact that after 80 min. the impact value was only 27 ft.lb. (*i.e.*, partially brittle) implies that some embrittlement does take place at 620° C. It becomes clear that this is the reason for the intergranular fracture of steel *B* in the unembrittled condition. The intergranular nature of the fracture, and the fact that embrittlement occurs at 525° C. and is at least partly reversible, lead to the conclusion that the phenomenon concerned is temper-brittleness in the usual sense of the term, but in a much more severe form.

Table VI

ROOM-TEMPERATURE IMPACT VALUES OF STEEL *B*

Initial Treatment	Time at 620° C., min.	Izod Value, ft.lb.
O.Q. from 900° C. T. for 2 hr. at 620° C. Embrittled for 4 hr. at 525° C.	0	2
	5	4
	10	5
	20	13
	40	30
	80	27

MECHANICAL IMPLICATIONS OF TEMPER-BRITTLENESS

To understand clearly the factors influencing the measurement of temper-brittleness, the method of testing must be considered in some detail. It is now well established that the effect of the transformation is to raise the transition temperature in a notched-bar impact test, and it is also clear that the normal tensile properties are little affected in most cases. By applying Orowan's theory of notch brittleness,[11] it is easy to see the significance of these observations.

If the mean constraint factor k is defined as the ratio of the mean axial yield tension in the metal under a notch in a stressed bar to the uniaxial yield stress Y, it has been shown[12] that this factor cannot rise above about 3 in ductile materials. It follows, therefore, that no crack or notch can raise the maximum tensile stress above $3Y$, and a material can only show notch brittleness if its brittle strength B, at a stage of plastic deformation, becomes less than about $3Y$.

The effect of temperature of testing on the notched-bar impact value of a metal depends therefore on the relative values of B and Y, and their dependence on temperature. In general terms, the transition temperature occurs when $B = kY$.

It is useful to find equations to express the temperature dependence of Y and B on temperature. A simple relationship that may be used is of the form

$$Y = Ae^{-bT} \quad \text{.....(1)}$$

and it is assumed that

$$B = Ce^{-dT} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (2)$$

where A, b, C, and d are constants depending on the material, and T is the absolute temperature.

Then, at the transition temperature T, $kY = B$, where k is the value of the mean constraint factor appropriate to the conditions;

$$\therefore kAe^{-bT} = Ce^{-dT}$$

$$\therefore T = \frac{1}{(d-b)} \log_e \frac{C}{kA} \quad \ldots\ldots\ldots\ldots\ldots\ldots (3)$$

Values for A and b are readily available for a wide range of steels. Values for C and d are scarce, but values for Armco iron have been given by de Haas and Hadfield.[13] These indicate that $b \simeq 10d$, and is probably less for alloy steels. Since, in general, values for d are not known, it will be neglected in comparison with b in this treatment. Equation (3) then becomes

$$T = \frac{1}{b} \log_e \frac{kA}{C} \quad \ldots\ldots\ldots\ldots\ldots\ldots\ldots\ldots (4)$$

i.e., the transition temperature is a function of the logarithm of the brittle strength. This simple equation depends on the form of the relationship given in equation (1). The relationship in equation (1) is not greatly in error in the range 80–380° K.

When comparing the transition temperatures in notched-bar tests and in tensile tests, the difference between these two values (ΔT) is given by the expression

$$\Delta T = \frac{1}{b} \log_e k$$

where k is the value of the mean constraint factor appropriate to the notched-bar test. For a sharp notch, $k = 2{\cdot}5 - 3$, *i.e.*, $\log_e k \simeq 1$. Now $1/b$ varies from 120 for pure iron to about 700 for quenched and slightly tempered alloy steels. It is clear, therefore, that for a series of steels of the same transition temperature in the notched-bar test, the low-temperature tensile properties will depend almost completely on the temperature dependence of the yield stress.

Thus, pearlitic steels ($1/b \simeq 400$), which in general have higher transition temperatures than pure iron ($1/b \simeq 120$), are moderately ductile in liquid air, whereas pure iron is very brittle.

A further point of interest is that, for carbon steels, the slope of the ductility/temperature curves in the transition range seems to correlate with $1/b$. Rinebolt and Harris[14] have given a series of transition curves which show that as the carbon content of pearlitic steels is increased, the curves become less steep. Data given by Teed[15] indicate that $1/b$ increases by a factor of 3 in the range of carbon content considered. However, the effect of other alloying elements is more complex. Additions of manganese to pure iron increase $1/b$ without altering the sharpness of the transition during the first fall in impact value. The curve becomes less steep in the lower range of the transition.[16]

When steels are temper-embrittled, it is known that:

(i) The transition temperature is raised,
(ii) The yield stress is unaltered, and
(iii) The dependence of yield stress on temperature is unaltered (deduced from Wiester,[17] who carried out low-temperature tensile tests on steels that were almost certainly temper-brittle).

Therefore A and b are unaltered by the transformation, and equation (4) can be written

$$\text{Transition temp.} = \text{const.} - \text{const.} \times \log_e C$$

where C is the brittle strength.

The change in transition temperature ΔT, due to an embrittling treatment, can be written

$$\Delta T = \frac{1}{b} \log_e \frac{(C_1)}{(C_2)}$$

where C_1 and C_2 are the brittle strengths before and after embrittlement, respectively.

Taking displacements of transition curves as a measure of the amount of embrittlement due to a given heat treatment is merely an indirect method of measuring changes in brittle strength. But ΔT also depends on $1/b$, *i.e.*, ΔT will vary greatly with the initial heat treatment of the steel. Light-tempered steels with $1/b \simeq 600$ should give larger values of ΔT than fully tempered steels. Pearlitic steels would also be expected to give smaller values of ΔT than steels with tempered martensitic structures, for the same reason. When determining the effect of alloying elements on temper-brittleness, any changes in $1/b$ must be allowed for.

Table VII

EFFECT OF AUSTENITIC GRAIN SIZE ON IMPACT PROPERTIES OF STEELS *S* AND *V*

Designation (En.23)	Hardness, V.D.H.		A.S.T.M. Grain Size	Transition Temperature, °C.	Maximum Impact Value, ft. lb.	Type of Fracture
	O.Q.	As Tested				
S	608	289	1–3	− 86	59	Cleavage and intergranular
SB	608	286	1–3	+ 34	58	Intergranular
SB250	608	278	1–3	+ 46	58	Intergranular
S	609	284	8–9	−112	65	Cleavage
SB	609	281	8–9	− 42	65	Intergranular
V	604	298	1–3	+ 30	56	Mostly intergranular
VB	604	299	1–3	+206	43	Intergranular
VB250	604	289	1–3	+230	33	Intergranular
V	596	291	7–8	− 78	66	Mostly cleavage
VB	596	289	7–8	+ 46	64	Intergranular

Steel *S* contains 0·021% As; Steel *V* contains 0·255% As

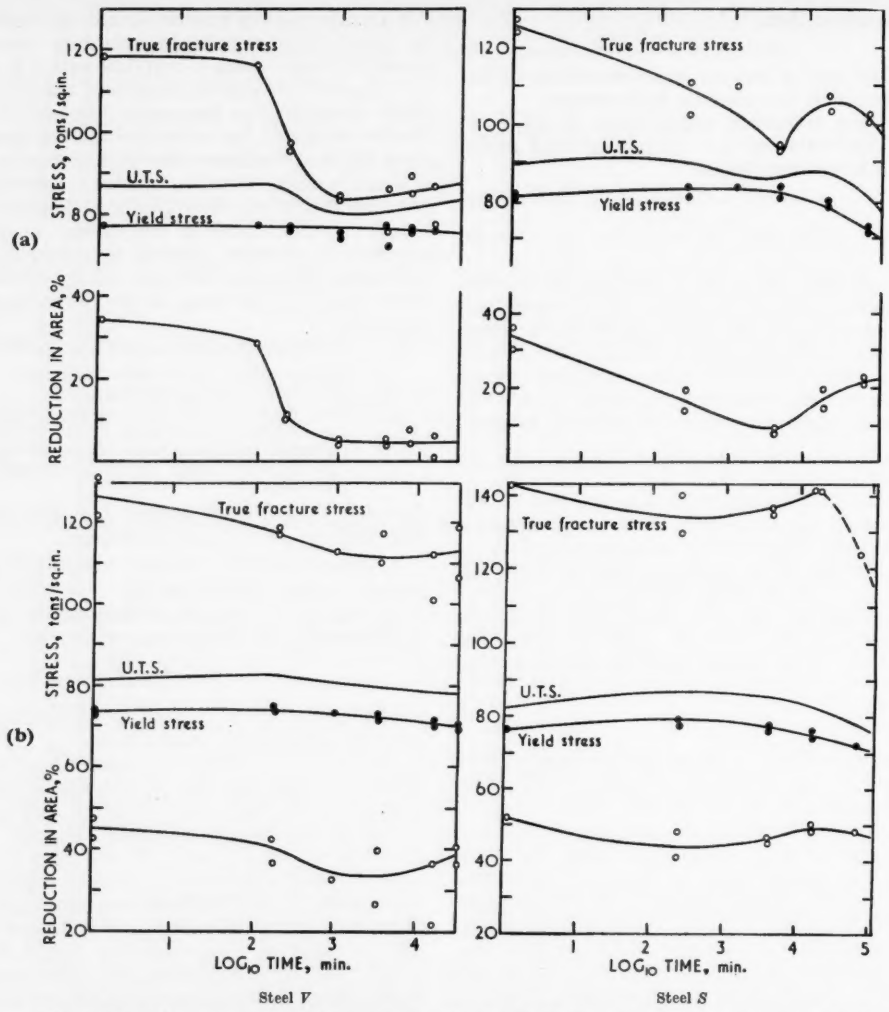


Fig. 3—Tensile properties of (*a*) coarse-grained and (*b*) fine-grained specimens of steels *V* and *S*, tested at − 196° C. Both steels were O.Q. from (*a*) 1150° C. or (*b*) 900° C.; tempered for 2 hr. at 620° C.; W.Q.; and heated at 525° C. for the times indicated

From this reasoning, it can be deduced that steels that have been severely temper-embrittled should have markedly low ductility in liquid air. The results of experiments designed to investigate this are given in the next section.

In the general development, it has been tacitly assumed that only one type of brittle failure and one brittle strength is involved. In practice, there are two, cleavage and intergranular, corresponding to the unembrittled and embrittled conditions respectively. For the theory to be made applicable, it must be assumed that both types of failure have the same criterion for fracture (*e.g.*, critical normal stress).

EFFECT OF EMBRITTLEMENT ON LOW-TEMPERATURE TENSILE BRITTLENESS

Five series of tests were carried out to investigate the effect of embrittlement on the low-temperature tensile properties. Two of the arsenic-bearing steels, each with a coarse and fine austenitic grain size, and one of the antimony steels (fine-grained) were tested.

A series of specimens of the En.23 steels *V* and *S* (0·255% and 0·021% As) were austenitized at 1150° C. for 90 min. and oil-quenched. This produced an austenitic A.S.T.M. grain size of 1–3. These were tempered for 2 hr. at 620° C., and part of each batch was embrittled at 525° C. for 4 hr. or 250 hr. The transition curves were determined, and the relevant data are given in Table VII, which also gives the equivalent data for the fine-grained steels. The susceptibility is approximately doubled by the increase in grain size, and the effect of increasing the arsenic content by a factor of 10 is about equivalent to increasing the A.S.T.M. grain size from 8–9 to 1–3. The transitions were found to be more gradual in the coarse-grained steels.

Low-temperature tensile test pieces were made from these fractured impact specimens and from some of the unembrittled specimens that had been embrittled for various times at 525° C. in the range 1–500 hr. The test pieces were then fractured at − 196° C. The results of these tests are shown in Fig. 3*a*.

Figure 3*a* for the coarse-grained high-arsenic steel *V* shows that as the time at 525° C. is increased, the true stress at fracture and the reduction in area decrease very rapidly at about 100–200 min., corresponding to a transition temperature of about + 200° C. in the impact tests. Thereafter, only small changes occur. The yield stress remains nearly constant, but the ultimate tensile strength falls after 100 min. This drop takes place when fracture occurs without local reduction in area. The yield-points in all the low-temperature tests were very prominent, but were only just detectable in room-temperature tests. For the most brittle specimen, fracture occurred immediately after yielding, but before the load had started to rise again.

The specimens with low ductility fractured at stresses between 80 and 90 tons/sq. in., and it is certain that these values approach the true values of the brittle strengths for these specimens. By prestraining a specimen at room temperature and subsequently fracturing it in liquid nitrogen, it was possible to get a rough estimate of the effect of deformation on the brittle strength. This was used to correct for deformations in the more brittle specimens. That embrittled for 250 hr. at 525° C. was found to have a corrected brittle strength of 78 tons/sq. in. This corresponds to a notched-bar transition temperature of + 230° C.

The fractures of the tensile specimens showed interesting changes as the embrittlement proceeded. The unembrittled steel gave a flat transverse fracture with no visible shear; many fine cracks were visible on the surface. After 90 min. at 525° C., the low-temperature fracture was recognizably intergranular, but the surfaces of the grains were dull in appearance. After 120 hr., the grain surfaces were bright, typical of the low-temperature impact fracture. The room-temperature tensile fracture of this steel also appeared to be intergranular, and had a noticeably low ductility (43% reduction in area).

The general conclusion to be drawn is that the effect of embrittlement at 525° C. on the low-temperature load-extension curves is to cause a premature arrest, at successively lower extensions, as the embrittlement proceeds. The changes in ductility are accompanied by significant changes in the appearance of the fracture.

Figure 3*a* also shows the results for the low-arsenic coarse-grained steel. Two new effects are immediately apparent from the curves. Both the true fracture stress and the reduction in area fall with increasing embrittling time up to about 60 hr. The U.T.S. also shows a slight fall at this time, indicating failure without necking. Both the fracture stress and the ductility increase at longer embrittling times. This seems to be due to the 'over-ageing' reported by Vidal.[18] At longer times, the true fracture stress falls again slightly. This is accompanied by a fall in yield stress, U.T.S., and hardness, and must therefore be due to softening of the material. The fall in ductility of the unembrittled material on changing the testing temperature from 20° to − 196° C. was from 57% to 33·5% reduction in area, similar to that for steel *V*, indicating that arsenic has little effect on this property. The room-temperature ductility of the specimens given 1000 hr. at 525° C. was nearly the same as that of the unembrittled steel, but the yield stress had fallen considerably. The lower susceptibility of this steel is therefore apparent in tests both at room temperature and at − 196° C.

Series of tests were also carried out on the high- and low-arsenic steels in the fine-grained condition. Specimens were taken from the fracture-impact specimens, and also from some of the unembrittled pieces that had undergone prolonged embrittlement at 525° C. (up to 1000 hr.). The results of these tests are shown in Fig. 3*b*.

The fine-grained steel *V* (0·255% As) in Fig. 3*b* shows effects similar to the coarse-grained series, but in a much less pronounced form. It is clear that the fracture stress and ductility fall with time up to 60 hr. The values at longer times are scattered, but show some tendency to decreased brittleness. All the test pieces in this series showed local reduction in area. The slight fall in U.T.S. is due to softening. The fall in ductility of the unembrittled material caused by changing the testing temperature from 20° to − 196° C. (from 64% to 47·5% reduction in area) was relatively small, and the minimum reduction in area was only 30% (apart from one very scattered point). The fractures of these specimens differed only slightly from one to another, and in no case could they be identified as intergranular. These specimens were particularly interesting, since they gave clear evidence of the effect of temper-brittleness on fracture at large deformations.

The fine-grained low-arsenic steel *S* in Fig. 3*b* showed even less response to the embrittling treatment. A slight reduction in ductility and fracture stress took place at the shorter times; the steel recovered to nearly the original value after 250 hr. At longer times, the softening became appreciable and the yield stress, U.T.S., and fracture stress all fell slightly, without further lowering of the reduction in area. The initial changes in ductility and fracture stress are probably associated with embrittlement and subsequent 'over-ageing,' as already mentioned.

These four sets of results show that the austenitic grain size was a more important factor than the arsenic content in influencing the low-temperature ductility; this applies to both the embrittled and unembrittled steels. The low-temperature ductility shown by the unembrittled steels gives a good indication of the response to the embrittling treatments. When this ductility is low, extreme brittleness can be produced by heat treatment at 525° C. When the ductility was sufficiently low, the fractures could be identified as intergranular (*i.e.*, if the reduction in area was less than 30%).

The stress at fracture was reduced by temper-embrittlement (this might be better expressed as 'the stress at fracture depends on the ductility'). For the most brittle specimens, the fracture stress was about 80 tons/sq. in.

XUM

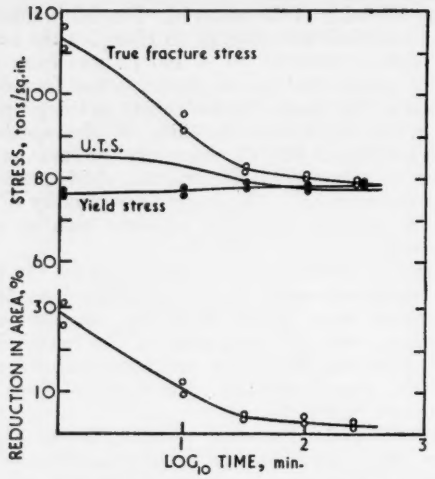


Fig. 4—Tensile properties of steel *B* (En.23 + 0·085% Sb), tested at − 196° C. Heat treatment as in Fig. 3*b*

These effects can be interpreted in the following way. Temper-embrittlement reduces the intergranular brittle strength of low-alloy steels. Only severe embrittlement reduces it to the value of the yield stress of the undeformed material. In other cases, deformation takes place before fracture, until the instantaneous value of the yield stress is increased to that of the brittle strength, when fracture takes place. It is not clear whether this type of mechanism applies to those specimens with reduced ductility that did not give recognizably intergranular fractures.

For the steels *A* and *B* containing antimony, the effects of embrittlement at 525° C. were very marked. The results of the low-temperature tensile tests are shown in Fig. 4. The test pieces were made from the fractured impact specimens referred to earlier, some of which were given additional embrittling treatments. The true stress at fracture and the reduction in area fall precipitously, as the time at 525° C. increases, to about 80 tons/sq. in. and 1% respectively. A new effect is also evident; a slight but definite increase took place in the yield stress and hardness as the embrittlement increased. All the low-temperature fractures, except those of the unembrittled specimens, were found to be intergranular and, as already mentioned, the room-temperature fracture after 4 hr. at 525° C. was also found to be so. The changes in properties of this steel are seen to be similar to those in the arsenic-bearing steels, again confirming the proposal that the phenomenon dealt with is a typical case of temper-brittleness, in a severe form.

By using the results for steel *V* (0·255% As), it is possible to link the tensile and notched-bar tests for the rest of the series. For this steel, in the coarse-grained condition, the brittle strength after 250 hr. at 525° C. was found to be 78 tons/sq. in., and the transition temperature in the Izod test was + 230° C. Assuming that the residual ductility has been adequately allowed for, by substituting in equation (4) it is found that $k = 2{\cdot}4$. By using this value of k and the values of A and b obtained from the yield-stress determinations, it is possible to obtain the brittle strengths corresponding to the notched-bar transition temperatures (*see* Table VIII). These figures are very tentative, but they show the magnitude of the changes taking place, and emphasize the real influence of arsenic and antimony on temper-brittleness. The changes in brittle strength are quite large. The wide range of values (153–76·5 tons/sq. in.) implies that substantial effects are taking place, even though it is only a grain-boundary phenomenon.

By inserting the values of $1/b$ and k in equation (4), $T \simeq 425°$ K. for the En.23 steels. This means that for complete brittleness in tension in liquid air, the transition temperature in the impact test must be above about 240° C. It can be seen from this that, in general, this extreme brittleness does not arise, but an appreciable fall in the low-temperature ductility takes place when the transition temperature rises above room temperature. This is a consequence of the relatively gradual lowering of the ductility as the temperature is lowered in the tensile test.

Experiments Concerning the Mechanism of Embrittlement

One of the most striking features of temper-brittleness is that the effect is intergranular with respect to the prior austenitic grain boundaries. The fractures of temper-brittle steels are intergranular,[9, 19] and, when etched in suitable reagents, the austenitic grain boundaries are heavily attacked.[20, 21] It may also be noted that the grain-boundary carbides are left more or less unattacked by these reagents.

Table VIII

BRITTLE STRENGTHS CORRESPONDING TO NOTCHED-BAR TRANSITION TEMPERATURES

Steel	A.S.T.M. Grain Size	T_a*	C_a†	T_b*	C_b†	T'_b‡	C'_b‡	ΔC†	ΔT*
En.23, 0·255% As	1–3	+ 30	119·6	+206	82·4	+230	78	37·2	176
En.23, 0·255% As	7–8	− 78	136·4	+ 46	104·5	...	...	31·9	124
En.23, 0·022% As	1–3	− 86	149	+ 34	112·6	+ 46	109·4	36·4	120
En.23, 0·022% As	8–9	−112	153·6	− 42	126·1	...	...	27·5	70
En.23, 0·088% Sb	7–8	− 24	133·8	+293**	76·5	...	...	57·3	317**

* T_a and T_b are the unembrittled and embrittled transition temperatures, ° C.; $\Delta T = T_a - T_b$
† C_a and C_b " " " " " brittle strengths, tons/sq. in.; $\Delta C = C_a - C_b$
‡ Refer to steels embrittled for 250 hr.
** Calculated on the assumption that $C_b = 76{\cdot}5$ tons/sq. in.

(a) (b) (c)

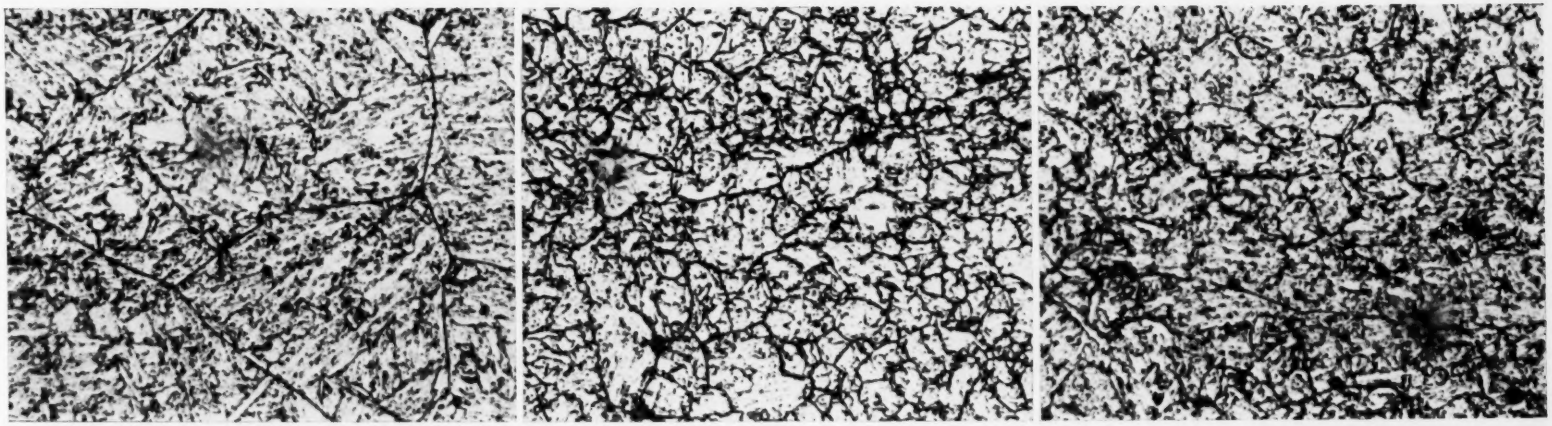


Fig. 5—Structure of steel *T* (0·088% As). All specimens O.Q. from 1200° C., given intermediate heat-treatment, and O.Q. again ; then T. at 625° C. and embrittled at 525° C. Etched in 1% CTAB : (*a*) No intermediate heat-treatment ; (*b*) 16 min. at 800° C. ; (*c*) 2 min. at 850° C. × 650

(a) (b) (c)

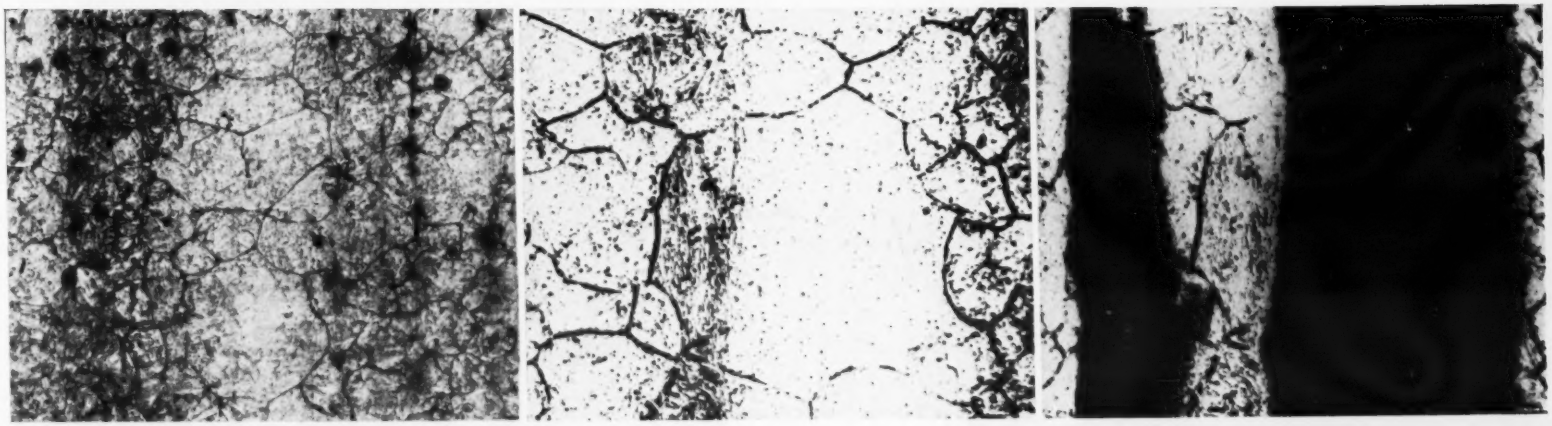


Fig. 6—Longitudinal section of embrittled En. 23, O.Q. from 1150° C., T. at 620° C., and embrittled at 525° C. : (*a*) Etched in 1% CTAB, × 120 ; (*b*) as (*a*), × 650 ; (*c*) as (*b*), etched further with Stead's reagent

Experiments on steel *T* have shown that the grain-boundary etching can persist from one austenitizing treatment to the next when the second austenitizing treatment is at a sufficiently low temperature.

A series of specimens of steel *T* (En.23, 0·088% As) were heated at 1200° C. for 1 hr. and oil-quenched. Some of these were then heated in a salt bath at 900° C. for various times, followed by oil-quenching. The specimens were next tempered at 620° C. and embrittled for 24 hr. at 525° C. The procedure was repeated with intermediate austenitizing temperatures of 800° and 850° C. on other specimens.

Figure 5*a* shows the structure of the steel when the intermediate austenitizing treatment is omitted. These steels were all etched in an ethereal solution of picric acid to which 1% CTAB had been added.[21] Figure 5*b* shows the structure with an intermediate treatment of 16 min. at 800° C. Two sets of grain boundaries are clearly visible, one of which corresponds to the original coarse grains. Figure 5*c* shows an area after an intermediate treatment of 2 min. at 850° C., with an old grain boundary clearly visible. After 30 min. at 850° C., very little trace of these boundaries remained. At 800° C., the corresponding time was about 2 hr.; after 2 min. at 900° C., no clear evidence of the original grain-boundary structure was found. These experiments show that the effects of the austenitic grain boundaries can persist through an austenitizing treatment, or, more precisely, that etching occurs at the sites of the old grain boundaries. This indicates that segregation of material in solution to the austenitic grain boundaries may be responsible for embrittlement. This segregation, if it occurs when the steel is austenitic, would take an appreciable time to disperse when given the intermediate heat treatment. Persistence of the etching would be expected to a degree depending on the temperature and time of the intermediate austenitizing treatment. It is interesting to note that fracture may take place along the old grain boundaries when the steel is suitably treated.

It seems to be possible that temper-brittleness is dependent on grain-boundary segregation while the steel is austenitic, and that tempering in the embrittling range leads to changes in the segregate, which actually causes the brittleness. This has been proposed by Spretnak and Speiser,[22] but the details of their mechanism seem to be improbable.

Any treatment that might lessen this segregation should obviously reduce the susceptibility. Jones[23] demonstrated that prolonged tempering at 650° C. reduced the susceptibility as measured by room-temperature tests, but this may have been a spurious effect. Measurements of transition temperatures of steels given prolonged tempering treatments, followed by embrittlement, have been made.

A series of specimens of steels *T* and *V* (En.23, 0·085% and 0·255% As) were oil-quenched after 1 hr. at 900° C. These were then tempered at 620° C. for 30 min., 8 hr., 32 hr., or 128 hr., and water-quenched. Half of each batch was then embrittled for 4 hr. at 525° C., followed by water-quenching. The transition curves were determined for the whole series. Figure 7 shows the change in transition temperature due to the embrittling treatment as a function of hardness;

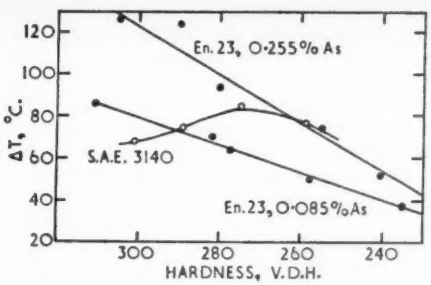


Fig. 7—Effect of hardness after tempering at 620° C. on the subsequent embrittlement at 525° C. The abscissa gives the hardness after tempering at 620° C. for varying times (proportional to log time at 620° C.), and the ordinate gives the change in transition temperature due to the subsequent treatment of 4 hr. at 525° C.

the hardness is proportional to the logarithm of the time at 620° C. It is clear that the susceptibility is much reduced by prolonging the tempering, and this takes place largely by the lowering of the embrittled transition temperature. The unembrittled transition temperature is hardly affected by increasing the tempering time above 2 hr.

Also plotted on this graph are some of the results of Buffum and Jaffe[24] for an S.A.E. steel 3140 (0·77% chromium, 1·26% nickel, and 0·39% carbon). These workers obtained different hardnesses by using different tempering temperatures in the range 500–675° C., and followed this with embrittlement at 450° C. They concluded that susceptibility was hardly affected by altering the hardness in this way. Their results were probably affected by embrittlement taking place at the lower tempering temperatures before the specific embrittling treatment used. The results obtained in the present work confirm the hypothesis of grain-boundary segregation.

The precise mechanism of the changes at the austenite grain boundaries during embrittlement is more obscure. Manganese, chromium, and nickel, which increase susceptibility when present in substantial amounts, seem to act differently from phosphorus, arsenic, and antimony, which are effective in much smaller amounts. Molybdenum reduces susceptibility, but carbon–molybdenum steels (2·1% Mo) are quite susceptible. The effect of carbon on susceptibility is not clear. The action of molybdenum is very significant, since this element is known to decrease the banding, or macro-segregation, in manganese steels.[25] It is possible that it also decreases the grain-boundary segregation, and so lowers the susceptibility.

Certain interesting etching effects were found to be associated with the banding in the nickel-chrome steels studied. When longitudinal sections of temper-brittle specimens were etched in CTAB, it was found that light etching areas of coarse austenite grains were associated with the banding. Figure 6*a* shows a typical area of a coarse-grained specimen (O.Q. from 1150° C., tempered at 620° C., embrittled 4 hr. at 525° C.). An area of a similar specimen at a higher magnification is shown in Fig. 6*b*. When etched with

XUM

Stead's reagent, the same area appeared as shown in Fig. 6c. Comparison shows that copper has been deposited on the light etching areas. The dark etching areas are therefore areas rich in phosphorus, arsenic, and antimony. As the grain-boundary etching is more prominent in these areas, it seems safe to conclude that these elements are closely associated with the grain-boundary etching, and hence with the embrittlement.

In this way, it is possible to relate phosphorus, arsenic, and antimony with the embrittlement by means other than mechanical tests.

Since manganese, nickel, and chromium are also involved (this is deduced from mechanical tests), it does not seem possible to conclude that one particular compound or reaction is solely responsible for the embrittlement. Future work must be directed to a precise identification of the grain-boundary material. Both micro-radiography and auto-radiography will be useful in this respect, provided that the resolving powers of these techniques can be improved somewhat. Micro-analysis of material electrolytically extracted from fracture surfaces is another promising field of research.

GENERAL CONCLUSIONS

1. Temper-brittleness can be regarded as a manifestation of the lowering of the grain-boundary brittle strength. Arsenic slightly, and antimony greatly, increase the susceptibility of nickel-chrome steels; they also raise the transition temperatures of these steels in the absence of temper-brittleness.

2. The low-temperature tensile properties can show marked changes due to temper-embrittlement, but these changes are only slight in some circumstances, depending on the precise composition and heat treatment. When embrittlement is sufficiently severe, the room-temperature tensile properties are impaired.

3. Arsenic has little influence on the normal tensile properties, but antimony lowers the ductility, because it makes the steel so prone to temper-brittleness.

4. Increasing the time of tempering at 625° C. produces a progressive decrease in susceptibility to temper-brittleness.

5. The degree of grain-boundary etching is associated with the banding present in the steels studied. It is concluded that phosphorus, arsenic, and antimony are closely associated with the etching and the brittleness.

6. It is suggested, in support of earlier findings, that temper-brittleness is due to the segregation of solute atoms to the grain boundaries while the steel is austenitic. These segregations are retained to a greater or lesser extent when the steel cools to room temperature. Subsequent heat treatment in the range 300–600° C. leads to changes in the grain-boundary structure which result in the brittleness. The nature of the grain-boundary material is not known. It is suggested that some attempt should be made to identify specifically the grain-boundary elements.

Acknowledgments

The authors would like to express their thanks to the British Iron and Steel Research Association for their financial support, which included the award of a bursary to one of the authors (A. R. E.). They also wish to thank Mr. H. Bull of Brown Bayley Steels, Ltd., for his co-operation and generosity in making available the steels for the investigation and arranging for their analysis, and Dr. J. Nutting of the Department of Metallurgy, Cambridge University, for many helpful discussions.

References

1. H. Jolivet and G. Vidal: *Rev. Mét.*, 1944, vol. 41, pp. 378–388 and 403–408.
2. M. Baeyertz, W. F. Craig, jun., and J. P. Sheehan: *J. Met.*, 1949, vol. 1, pp. 535–543.
3. M. Baeyertz, W. F. Craig, jun., and J. P. Sheehan: *Ibid.*, 1950, vol. 188, pp. 389–396.
4. E. Houdremont, H. Bennek, and H. Neumeister: *Arch. Eisenhüttenwesen*, 1938, vol. 12, pp. 91–101.
5. B. Jones and J. D. D. Morgan: *J. Iron Steel Inst.*, 1939, No. II, pp. 115–136.
6. Second Report of the Alloy Steels Research Committee, Spec. Rep. No. 24, p. 217: 1939, London, The Iron and Steel Institute.
7. J. D. Sullivan and R. A. Witschey: *Metals and Alloys*, 1936, vol. 7, pp. 171–173.
8. G. Hagg: *Nova Acta Regiae Soc. Sci. Upsaliensis*, 1929, vol. 7, pp. 1–96.
9. A. R. Entwisle: *J. Iron Steel Inst.*, 1951, vol. 169, pp. 36–38.
10. J. H. Hollomon: *Trans. Amer. Soc. Metals*, 1946, vol. 36, pp. 473–540.
11. E. Orowan: "Reports on Progress in Physics," vol. 12, pp. 185–232: 1948–49, London, The Physical Society.
12. E. Orowan, J. F. Nye, and W. J. Cairns: Theoretical Research Report No. 16/45, Armament Research Dept., Ministry of Supply: 1945, London, H.M.S.O.
13. W. H. de Haas and R. A. Hadfield: *Phil. Trans. Roy. Soc.*, Ser. A, 1933, vol. 232, pp. 297–332.
14. J. A. Rinebolt and W. J. Harris, jun.: *Trans. Amer. Soc. Metals*, 1951, vol. 43, pp. 1175–1201.
15. P. L. Teed: "The Properties of Metallic Materials at Low Temperatures," vol. I: 1950, London, Chapman and Hall.
16. W. P. Rees, B. E. Hopkins, and H. R. Tipler: *J. Iron Steel Inst.*, 1951, vol. 169, pp. 157–168.
17. H.-J. Wiester: *Stahl u. Eisen*, 1943, vol. 63, pp. 41–47 and pp. 64–68.
18. G. Vidal: *Rev. Mét.*, 1945, vol. 42, pp. 149–155.
19. J. Nutting and V. E. Coslett: "Metallurgical Applications of the Electron Microscope," p. 61: 1950, London, Institute of Metals.
20. J. B. Cohen, A. Ulrich, and M. Jacobsen: *Trans. Amer. Soc. Metals*, 1947, vol. 39, pp. 109–136.
21. D. McLean and L. Northcott: *J. Iron Steel Inst.*, 1948, vol. 158, pp. 169–177.
22. J. W. Spretnak and R. Speiser: *Trans. Amer. Soc. Metals*, 1951, vol. 43, pp. 737–747.
23. J. A. Jones: *Metal Treatment and Drop Forging*, 1938, vol. 4, pp. 97–101.
24. D. C. Buffum and L. D. Jaffe: *J. Met.*, 1951, vol. 3, p. 540.
25. K. J. B. Wolfe and I. R. Robinson: *Metal Treatment and Drop Forging*, 1949, vol. 16, pp. 209–218.

Temper-Brittleness in High-Purity Iron-Base Alloys

By A. Preece, M.Sc., F.I.M., and R. D. Carter, Ph.D.

SYNOPSIS

The effect of composition on the occurrence of temper-brittleness has been examined by microscopical methods and measurement of transition temperatures.

By the criteria of rise in transition temperature and development of an intercrystalline brittle fracture after treatment at 500° C., Fe–C and Fe–C–P alloys were not found to be susceptible to temper-brittleness, but Fe–Cr–C–P and Fe–Mn–C–P alloys were highly susceptible. Reduction of carbon content to about 0·003% eliminated the embrittling change. Phosphorus was found to have a large and real effect in increasing susceptibility, but its presence may not be essential for the occurrence of temper-brittleness. Addition of nickel to alloys containing manganese or chromium appeared to increase the susceptibility, as judged by rise in transition temperature, but its effect is probably not one of direct increase in the extent of the embrittling change.

Picric acid etchants readily revealed the embrittling change at grain boundaries and sub-boundaries, but only in susceptible alloys that contained chromium. 725

THE majority of investigations on temper-brittleness have been carried out with commercial steels or modifications of such steels. In view of the limitations of these materials, due to the presence of numerous elements, it seemed desirable to investigate the phenomenon in simple iron-base alloys of high purity, in order to determine the elements responsible for temper-brittleness, and their particular characteristics in the embrittling process.

In early investigations the criterion of susceptibility to temper-brittleness was the fall in the notched-bar impact energy at room temperature. However, since the impact energy is dependent upon the temperature of testing, misleading conclusions may be drawn from impact-energy values at any single temperature, and determination of the temperature (or rather the range of temperature) at which there is a transition from tough to brittle fracture is necessary. In the present investigation, transition temperatures have been measured, and examination of the nature of the brittle fracture has shown that appreciable changes in transition temperature due to temper-embrittlement are invariably associated with intergranular facets in the brittle fracture.

The changes in microstructure revealed by certain picric acid etchants give a quite specific indication of the embrittling change, and have enabled it to be detected even where it had not proceeded to a sufficient extent to affect the transition temperature.

EXPERIMENTAL PROCEDURE

The alloys were prepared by vacuum melting, using a specially pure Swedish iron (B.I.S.R.A. reference *AAAID*) of the following composition:

C, %	Si, %	Mn, %	S, %	P, %	Ni, %
0·015	0·016	0·008	0·006	0·001	0·005
Cr, %	**Al, %**	**Cu, %**	**O_2, %**	**N_2, %**	**H_2, %**
0·0025	0·001	0·006	0·044	0·01	tr.

Alloy additions were made using spectrographically pure Mn and the purest commercial grades of ferro-chrome, ferro-manganese, and ferro-phosphorus.

The equipment available could not produce melts of more than about 600 g., and to carry out a sufficient number of transition-temperature determinations on each cast, a miniature impact test piece had to be used. As well as offering advantages in preparation and heat-treatment, the small test piece made possible the introduction or removal of carbon, nitrogen, or oxygen in reasonably short times. The alloys were melted in alumina crucibles and solidified under vacuum in the crucible. The resulting ingots were forged and rolled to $\frac{1}{2}$-in. dia. and were then drawn to $\frac{1}{8}$-in. dia. wire. The wire was cut into $\frac{3}{4}$-in. lengths to provide impact test pieces of the Hounsfield miniature size.

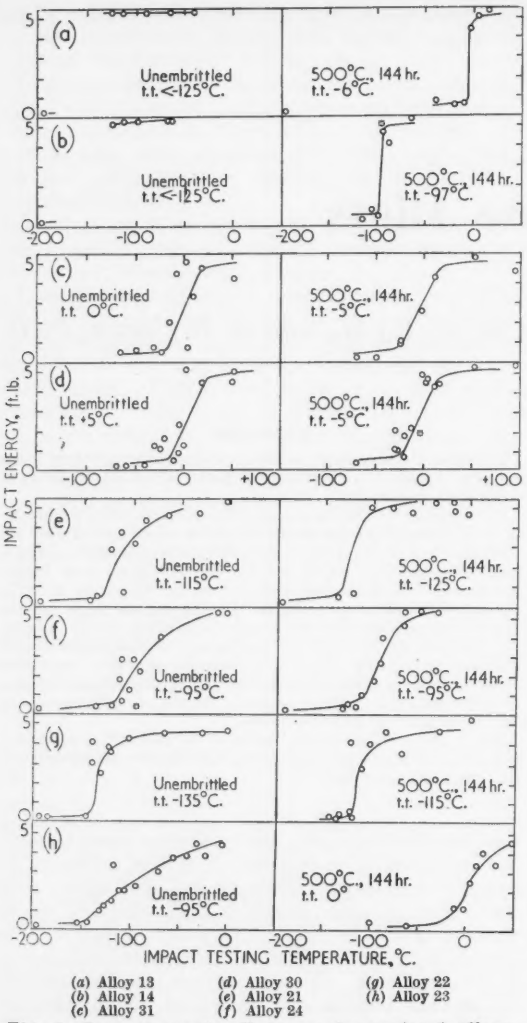
The transition temperatures obtained with these small specimens will be considerably lower than with normal Izod or Charpy test pieces.

Controlled-atmosphere treatments to modify carbon, nitrogen, and oxygen contents were carried out in the usual type of wire-wound Vitreosil tube furnace, with precautions to ensure adequate gas flow over the specimens.

Heat-treatment of the specimens was carried out *in vacuo* in a vertical tube furnace. The specimens were suspended in bundles on a wire that could be

Paper MG/A/250/52 of the Alloy Steels Research Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association, received 31st October, 1952. The views expressed are the authors' and are not necessarily endorsed by the Committee as a body.

Mr. Preece is Professor of Metallurgy at King's College, Newcastle-upon-Tyne. Dr. Carter, formerly a research student at King's College, is now a member of the research staff of British Insulated Callenders Cables, Ltd.



(*a*) Alloy 13 (*d*) Alloy 30 (*g*) Alloy 22
(*b*) Alloy 14 (*e*) Alloy 21 (*h*) Alloy 23
(*c*) Alloy 31 (*f*) Alloy 24

Fig. 1—Impact-energy diagrams for series 1 alloys

raised into the hot zone of the furnace without breaking the vacuum; the furnace was filled with argon before quenching. The specimens were notched after heat-treatment, so that any decarburization occurring during heat-treatment (this extended to a depth of not more than 0·001–0·002 in.) would not affect the material at the base of the notch.

The impact testing machine was designed to enable the miniature specimens to be tested over a wide range of temperature. The falling arm of the machine delivered a blow of 5·5 ft.lb. energy at a striking velocity of 9·6 ft./sec., and the specimen underwent beam-type loading as in the Charpy test. The test piece rested upon a steel anvil that could be immersed in a vacuum flask filled with the heating or cooling mixture, and when the required temperature had been attained the flask could be removed and the specimen broken within 1 sec. A chromel–constantan thermocouple situated about 0·1 in. below the notch of the test piece enabled temperatures to be measured with an accuracy of about 1° C. in the range — 196° to + 200° C.

The alloys as made were carbon-free, all carbon being added to the actual specimens by gas carburizing in an atmosphere that would introduce carbon to an equilibrium concentration. This was to ensure that with sufficiently long carburizing times the carbon content would rise to the desired value without exceeding it. A temperature of 950–960° C. was employed so as to obtain a single-phase structure during carburizing and to ensure rapid diffusion of carbon and attainment of equilibrium. The carburizing atmosphere consisted of hydrogen containing dipentene, as used by Wells and Mehl,[1] and was prepared by mixing two streams of dry hydrogen, one saturated at room temperature with dipentene and the other with no addition at all. By controlling the relative rates of flow of these two streams, the resulting carbon content of the alloys could be controlled over a wide range. The dry carburizing atmosphere was intended to ensure a low oxygen content in the carburized material.

Specimens were carburized in batches of up to 250 at a time for about one week. Control specimens of the alloys being carburized were rolled into strip (about 0·005 in. thick) and were included with the others in the carburizing tube for final carbon estimation by combustion. Micro-examination showed that carburization was quite homogeneous, both within individual specimens and from specimen to specimen. Notwithstanding the fine gauge of the test strip, the carbon content was the same as that of the round specimens, confirming that equilibrium had been reached. Carburization to the required composition was virtually complete in about half the time actually employed.

The specimens were given a standard pre-embrittling heat-treatment, *viz.*, water-quenched after 15 min. at 1000° C., then tempered for 1 hr. at 650° C. and water-quenched. This condition is subsequently referred to as the 'unembrittled' state, without implying that the embrittling change was completely absent in susceptible alloys.

First Series of Alloys

The compositions of the first series of alloys are shown in Table I; they were chosen so that the relative importance of Cr, Mn, Ni, and P, singly and in combination, could be determined. The lowest phosphorus content attainable was $< 0{\cdot}001\%$, and alloys containing this small amount are referred to as phosphorus-free.

Transition temperatures were determined in the unembrittled condition and after treatment at 500° C. for 144 hr. The results are shown in Table I, and Fig. 1 illustrates the consistency obtained in the impact tests.

A significant rise in transition temperature due to the embrittling treatment at 500° C. occurred only in the Fe–Mn–P–C and Fe–Cr–P–C alloys (14, 13, and 23), and then it was accompanied by the appearance of a large number of intercrystalline facets in the fracture. Figures 3 and 4 show typical fractures in the Fe–Cr–C and Fe–Cr–P–C alloys after treatment at 500° C. The zephiran-chloride/picric-acid etchant[2]

Table I

FIRST SERIES: COMPOSITION AND EMBRITTLEMENT TEST RESULTS

Alloy	Composition					Unembrittled		Embrittled 144 hr. at 500° C.	
	C, %	Mn, %	P, %	Ni, %	Cr, %	Trans. Temp., °C.	Brittle Fracture	Trans. Temp., °C.	Brittle Fracture
31	0·10	(N.P.L. deoxidized iron)				0	Transcryst.	−5	Transcryst.
30	0·12	...	0·036	...	...	+5	,,	−5	Transcryst.
14	0·10	1·4	0·010	...	...	Below −125	,,	−97	Mixed inter- and transcryst.
13	0·10	1·5	0·038	...	...	,,	,,	−6	Intercryst.
22	0·10	...	0·001	...	1·5	−135	,,	−115	Transcryst.
23	0·17	...	0·048	...	1·5	−95	,,	0	Intercryst.
21	0·10	...	0·001	1·4	...	−115	,,	−125	Transcryst.
24	0·17	...	0·048	1·7	...	−95	,,	−95	Transcryst.

showed clear differentiation between the unembrittled and embrittled states only in the Fe–Cr–P–C alloy (23). No difference was revealed in the microstructures of the two Mn-containing alloys, although there was a rise in transition temperature and an intercrystalline brittle fracture.

During tempering at 500° C., carbide precipitation was observed in all the alloys examined, both at grain boundaries and within the grains. The etching effect due to the temper-embrittling change was in no way connected with the grain-boundary carbides and appeared as a dark-etching region of the utmost observable fineness at grain boundaries and sub-boundaries. It was clearly evident in the embrittled state in alloys containing Cr and P, but almost unobservable in the unembrittled state. The Fe–Cr–C alloy (with only 0·001% of P), which showed no significant rise in transition temperature after treatment at 500° C., showed practically no dark-etching boundaries.

If a rise in transition temperature and the occurrence of intercrystalline facets in the brittle fracture is taken as a criterion for the occurrence of temper-brittleness, it appears that in iron-base alloys containing about 0·1% of C, the presence of 0·04% of P alone does not induce susceptibility, nor does 1·5% of Ni, with or without phosphorus; but Mn or Cr to the extent of about 1·5%, when 0·04% of P is also present, confer marked susceptibility. No temper-brittleness could be detected in the Fe–Cr–C alloy, and the results for the two Mn-containing alloys (13 and 14) suggest that no temper-brittleness would be detected in 1·5% Mn, 0·1% C alloys in the absence of phosphorus. Later experiments, however, showed slight temper-brittleness in alloys containing higher concentrations of Mn or Cr, despite a very low phosphorus content ($< 0·001\%$).

Second Series of Alloys

The eight alloys of the second series were prepared from the usual base materials and contained various proportions of Cr and phosphorus. They were intended to show the minimum concentration of Cr that will induce susceptibility to temper-brittleness in the presence of phosphorus.

The specimens were carburized to 0·1% of C and were prepared similarly to the first series, except that the embrittling treatment at 500° C. was for times varying from 10 to 670 hr.

Table II summarizes the compositions of the alloys and the transition temperatures before and after embrittlement; the right-hand column shows the difference between the transition temperature in the unembrittled condition and after heating for 670 hr. at 500° C., a positive value indicating that the transition temperature rose after embrittling treatment.

The brittle fractures of the first five alloys (54–58) were transcrystalline, both before and after embrittling. Alloys 61, 62, and 65 alone developed a marked

Table II

SECOND SERIES: COMPOSITION AND EMBRITTLEMENT TEST RESULTS

Alloy	Composition		Trans. Temp. (° C.) after Heating at 500° C. for—					Difference
	Cr, %	P, %	0 hr.	10 hr.	48 hr.	168 hr.	670 hr.	
54	0·80	0·044	−106		−93	−95	− 80	+ 26
55	0·91	0·013	−111	−102	−98	−96	− 90	+ 21
56	0·42	0·014	− 70			−86	− 80	− 10
57	0·50	0·050	−114	− 83	−94	−90	−100	+ 14
58	0·41	0·108	− 79			−74	− 58	+ 21
61	0·82	0·109	−112	− 50	−40	−23	− 10	+102
62	1·68	0·012	−122	− 80	−55	−44	− 37	+ 85
65	1·74	0·048	− 93	− 27	− 1	− 7	− 4	+ 89

Table III

THIRD SERIES : COMPOSITION

The elements added intentionally are in italics. Estimations of other elements (except As and Sb) were carried out spectrographically, and these values for impurities can be taken as typical for the method of preparation. The small amount of vanadium present in alloy 70 was added to enable it to be forged satisfactorily.

Alloy	C, %	Si, %	Mn, %	P, %	Ni, %	Cr, %	Mo, %	V, %	Cu, %	Co, %	Al, %	Sn, %	Sb, %	As, %
42	*0·20*	...	...	*0·045*	...	*2·40*	*0·51*	...	...	...	...	...	...	...
44	*0·19*	<0·01	*1·98*	<0·001	<0·01	0·01	<0·01	<0·01	0·01	<0·02	<0·005	<0·02	<0·005	<0·002
47	*0·19*	0·01	<0·01	<0·001	0·01	*2·30*	<0·01	<0·01	<0·01	<0·02	<0·005	<0·02	...	*0·062*
64	*0·20*	...	*0·98*	*0·048*	...	...	*0·47*	...	...	...	...	...	...	...
66	*0·19*	...	...	<0·001	...	*2·60*	...	...	...	...	...	...	<0·005	<0·002
67	*0·20*	...	...	<0·001	*4·00*	...	...	...	...	...	...	...	...	0·003
70	*0·19*	0·01	0·02	*0·119*	*3·80*	0·10	0·01	*0·10*	0·01	<0·02	<0·005	<0·02	...	...

intercrystalline brittle fracture after embrittling, but all the alloys of this series clearly showed dark-etching grain boundaries and sub-boundaries on embrittlement. The differentiation between the unembrittled and embrittled conditions was distinct in all cases, but those alloys giving the greatest rise in transition temperature showed the most marked differentiation.

When the Cr content was below approximately 0·8% and phosphorus 0·01%, the equilibrium temper-embrittling change was insufficient for intercrystalline brittle fracture or for a significant rise in transition temperature. With as little as 0·4% of Cr and 0·01% of P, the embrittling change was present but was so slight that it could be detected only by etching.

Previous investigators[3] have found that Mn steels show no susceptibility to temper-brittleness, as judged by the impact test, unless the Mn is above about 0·6%. The present results show that Cr behaves similarly, and it is likely that in Mn steels with less than 0·6% of Mn, the embrittling change also occurs but is not sufficient to cause any rise in transition temperature.

Third Series of Alloys

The compositions of the seven alloys of the third series are given in Table III. These alloys were examined to confirm earlier results which suggested that the presence of phosphorus might be essential for the development of temper-brittleness. They also provided for examination of the effect of Mo and As. The alloys were carburized to approximately 0·2% of C and were prepared as before. Figure 2 shows typical transition-temperature diagrams for alloy 44. Table IV shows the transition temperature and the nature of the brittle fracture for these alloys, both in the unembrittled condition and after 670 hr. at 500° C., and their response to the zephiran etchant. Hardness values are also included.

Results

Phosphorus-Free Alloys of Manganese and Chromium—The Fe–Mn–C alloy (44) and the Fe–Cr–C alloy (66) both showed a slight susceptibility to temper-brittleness, despite the very low phosphorus. The susceptibility of alloy 44 was such that the intergranular cohesion fell during embrittlement to a sufficiently low value to cause intercrystalline brittle fracture and a correspondingly marked rise in transition temperature. In alloy 66 the embrittling change was indicated best by the slight development of dark-etching boundaries after embrittling, and with long etching times (*i.e.*, in comparison with an unembrittled specimen etched similarly); the rise in transition temperature was quite small and was not accompanied by a marked amount of intercrystalline facets in the brittle fracture. Presumably, the intergranular cohesion was not lowered sufficiently by the embrittling process to lead to intercrystalline brittle fracture.

Phosphorus-Free Chromium Alloy, Containing Arsenic—The Fe–Cr–As–C alloy (47) was distinctly susceptible, showing a rise of 64° C. in transition temperature after embrittlement for 670 hr., accompanied by an appreciable amount of intercrystalline facets in the brittle fracture.

Influence of Molybdenum—Although both the Mo-containing alloys (42 and 64) of this series would have shown marked susceptibility in the absence of Mo, they showed very little when it was present. Although the rise in transition temperature was small in both cases, it seems that equilibrium is approached in much the same time as for Mo-free alloys, showing that Mo decreases the equilibrium degree of embrittlement rather than (as is often held) the rate at which equilibrium is attained. The possibility of any other explanation for the effect of Mo, other than that of a direct decrease in the amount of grain-boundary change, is ruled out by the etching evidence of alloy 42 (Fe–Cr–P–Mo–C), compared with that of alloy 40 (Fe–Cr–P–C), which is dealt with in the fourth series of alloys and shown in Figs. 5–8.

Nickel Alloys—The phosphorus-free Fe–Ni–C alloy (67) showed no susceptibility. The high-phosphorus alloy containing 3·8% of Ni (70) showed well-marked susceptibility, but as it also contained traces of Mn and Cr, its susceptibility might be regarded as due to the presence of Mn and Cr, together with a very

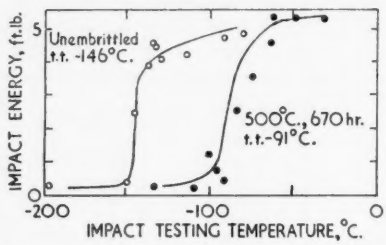


Fig. 2—Impact-energy diagrams for alloy 44 (Fe–Mn–C)

Fig. 3—Fe-Cr-C alloy (22), transcrystalline fracture

Fig. 4—Fe-Cr-P-C alloy (23), intercrystalline fracture

Figs. 3 and 4—Etched with zephiran-picric × 200

Fig. 5—Fe-Cr-P-C alloy (40), unembrittled

Fig. 6—Same alloy as in Fig. 5, after 670 hr. at 500° C.

Figs. 5 and 6—Same etching conditions (zephiran-picric) as Figs. 7 and 8 × 300

Fig. 7—Fe-Cr-P-Mo-C alloy (42), unembrittled

Fig. 8—Same alloy as in Fig. 7, after 670 hr. at 500° C.

Figs. 7 and 8—Same etching conditions (zephiran-picric) as Figs. 5 and 6 × 300

Fig. 9—Fe-Cr-P-C alloy (68), unembrittled

Fig. 10—Same alloy as in Fig. 9, after 670 hr. at 500° C.

Figs. 9 and 10—Same etching conditions (zephiran-picric) for both specimens × 200

Fig. 11 Fig. 12 Fig. 13 Fig. 14

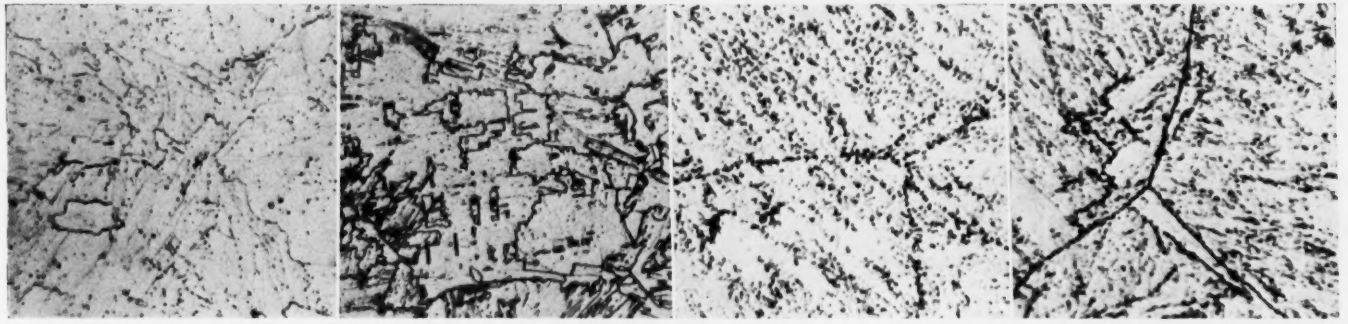


Fig. 11—Fe-Cr-P alloy (16), containing 0·014 % C, unembrittled
Fig. 12—Same alloy as in Fig. 11, after 110 hr. at 500° C.

Figs. 11 and 12—Same etching conditions (zephiran-picric) for both specimens × 700

Fig. 13—Fe-Cr-P-C alloy (16), unembrittled
Fig. 14—Same alloy as in Fig. 13, after 110 hr. at 500° C.

Figs. 13 and 14—Same etching conditions (aqueous picric acid) for both specimens × 2000

Fig. 15

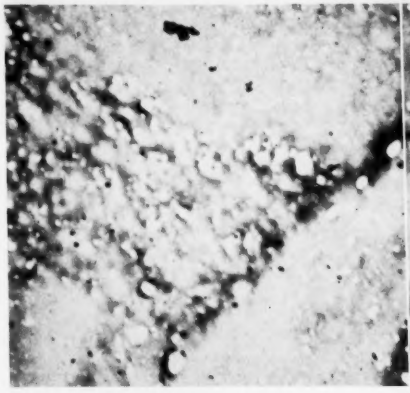


Fig. 16

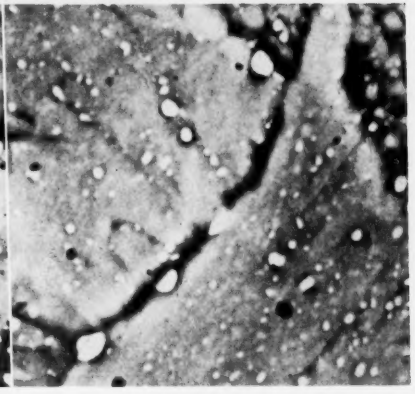


Fig. 15—Same specimen as in Fig. 13, electron micrograph
Fig. 16—Same specimen as in Fig. 14, electron micrograph

Figs. 15 and 16—Same etching conditions (alcoholic picric acid) for both specimens × 7500

Fig. 17

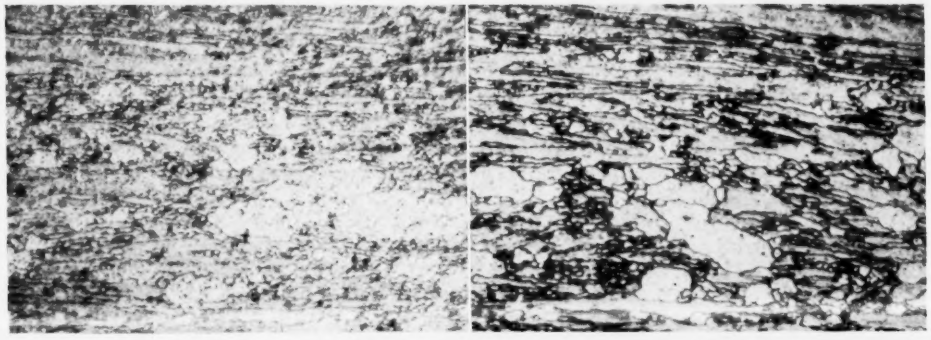


Fig. 18

Fig. 17—Cold-worked specimen after water-quenching from 650° C., showing partial recrystallization

Fig. 18—Same material as in Fig. 17, after treatment at 500° C., showing development of temper-brittleness

Figs. 17 and 18—Same etching conditions (zephiran-picric) for both specimens × 500

XUM

Table IV

THIRD SERIES : EMBRITTLEMENT TEST RESULTS

Alloy	Composition	Unembrittled			Embrittled 670 hr. at 500° C.			Response to the Zephiran Etchant
		Trans. Temp., ° C.	V.H.N.	Brittle Fracture	Trans. Temp., ° C.	V.H.N.	Brittle Fracture	
42	Fe–Cr–Mo–P–C	−121	243	Entirely transcryst.	− 95	252	Mainly trans-, possible tr. intercryst.	Slight
44	Fe–Mn–C	−146	180	,, ,,	− 91	168	Mainly trans-, tr. intercryst.	Nil
47	Fe–Cr–As–C	−123	227	,, ,,	− 59	209	Mainly trans-, some intercryst.	Nil
64	Fe–Mn–P–Mo–C	−137	241	,, ,,	−125	240	Entirely transcryst.	Nil
66	Fe–Cr–C	−105	229	,, ,,	− 85	223	Mainly trans-, possible tr. intercryst.	Slight
67	Fe–Ni–C	−108	176	,, ,,	−108	167	Entirely transcryst.	Nil
70	Fe–Ni–P–C (tr. Cr and Mn)	− 78	253	,, ,,	− 27	245	Mixed trans- and intercryst.	Nil

high phosphorus content, and made more evident by the presence of Ni. Although the results do not prove that a high-Ni, high-phosphorus alloy free from Mn and Cr is susceptible to temper-brittleness, they do show that any nominally high-phosphorus, high-Ni commercial steel is likely to be highly susceptible, and in this respect confirm the findings of Herres and Elsea.[4]

Fourth Series of Alloys

The alloys in this series (Table V) were highly susceptible to temper-brittleness and were examined to determine the influence of Mn, Cr, Ni, and P on the kinetics of the temper-embrittling process.

Specimens of the six alloys were carburized to approximately 0·2% of C, and were heat-treated as before, except that embrittlement was carried out at several temperatures in the range 450–625° C. for various times.

The isothermal embrittling characteristics of each alloy are shown in Figs. 19 and 20. The response to the zephiran etchant, the hardness, and the nature of the brittle fracture in the unembrittled state and after 670 hr. at 500° C., are shown in Table VI. Transition temperatures were taken as the temperature at which the impact energy had fallen to 1·5 ft.lb. This value was chosen instead of 2·5 ft.lb., as used for the first series, since the maximum impact-energy values for these materials with 0·2% of C were generally lower; also, since the fully tough impact-energy values showed an appreciable amount of scatter, the shape of the impact-energy/testing-temperature curve was better defined for the lower impact-energy values. The equilibrium extent of the embrittling change due to the pre-embrittling treatment at 650° C. was sufficient in some alloys, *i.e.*, 48, 60, and 68, to give rise to an intercrystalline brittle fracture. Figures 9 and 10 show the intercrystalline brittle fracture in alloy 68 in the unembrittled state and after 670 hr. at 500° C., and also the difference between the response to the zephiran etchant in these two conditions.

DISCUSSION OF RESULTS

Rate of Development of Temper-Brittleness

All the alloys appeared to have very similar isothermal embrittling properties, and it was difficult to find unique characteristics with either the Mn or the Cr alloys. At embrittling temperatures of 500° C. and above, however, approach to equilibrium was less rapid in the case of the two Mn alloys (45 and 49) than in the others, all of which contained Cr.

Composition and Degree of Embrittlement

Although the difference between the transition temperatures in the unembrittled and highly embrittled conditions is a useful guide to the susceptibility of a steel to temper-brittleness, it is not altogether satisfactory for measuring the real influence of each alloying element.

To interpret the present results a more satisfactory measure of susceptibility would be the difference between the intergranular cohesion of the alloy when

Table V

FOURTH SERIES : COMPOSITION

The elements added intentionally are shown in italics. With the exception of alloy 68, spectrographic analyses of these alloys gave the following results: Mo < 0·01%, Co < 0·02%, Al < 0·005%, Sn < 0·02%.
The preparation of alloy 68 was similar to the others and it would be expected to contain impurity elements in similar amounts. Nitrogen estimations made on several of the alloys gave values of < 0·001%.

Alloy	C, %	Si, %	Mn, %	P, %	Ni, %	Cr, %	V, %	Cu, %
40	*0·21*	0·01	<0·01	*0·044*	0·01	*2·40*	<0·01	<0·01
45	*0·19*	0·1	*1·36*	*0·048*	<0·01	0·01	<0·01	0·01
48	*0·19*	0·1	*0·99*	*0·053*	0·1	*2·31*	0·01	0·01
49	*0·20*	0·1	*1·60*	*0·045*	*3·0*	0·01	<0·01	0·01
60	*0·19*	0·05	0·12	*0·058*	*3.2*	*2·13*	0·01	0·01
68	*0·20*	0·01	0·01	*0·094*	0·01	*1·98*	0·01	0·01

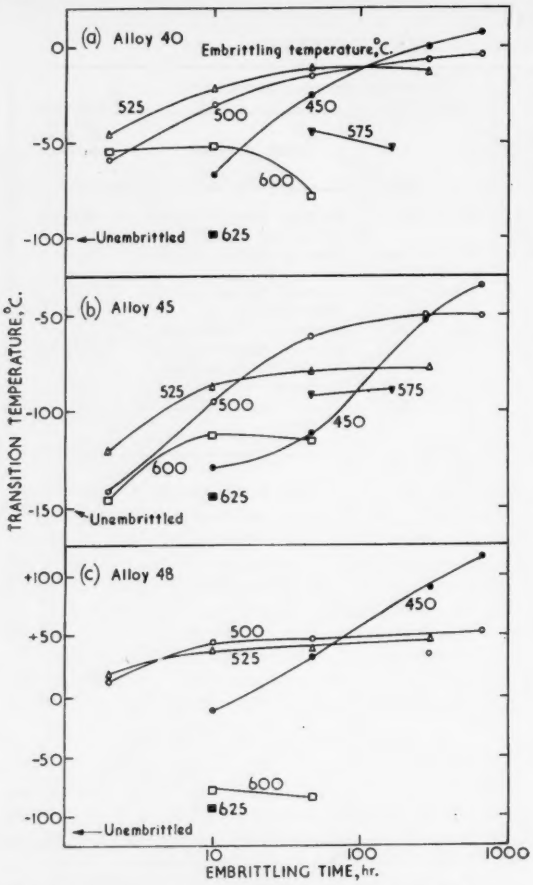


Fig. 19—Isothermal embrittling characteristics of alloys 40, 45, and 48 (series 4)

at equilibrium (with respect to the embrittling change), at 650°, and at 450° C. (conditions in these experiments of minimum and maximum embrittlement, respectively).

Where the brittle fracture of an alloy is intercrystalline, its transition temperature is a function of its intergranular cohesion, and it is reasonable to assume that the transition from tough to intercrystalline brittle fracture in the impact test occurs at that temperature at which the stress necessary to cause plastic flow of the ferritic matrix (under the specific stress distribution and strain rate of the test) slightly exceeds that required to produce rupture of the grain boundaries. Thus, in every impact test in which a transition from transcrystalline to intercrystalline brittle fracture occurs, the actual transition temperature* corresponds to a certain value of both ferritic yield stress and intergranular cohesion. At and below this temperature of transition from tough to brittle fracture, the fracture may be intercrystalline, transcrystalline, or a mixture of both. Those steels which give a predominantly intercrystalline brittle fracture in the unembrittled state enable the decrease in intergranular cohesion between the equilibrium states at 650° and 450° C. to be estimated in arbitrary units. This cannot be done for those which have a predominantly transcrystalline brittle fracture in the unembrittled state, for then the decrease in intergranular cohesion is clearly greater than comparison of the two transition temperatures would indicate, because a certain unknown degree of embrittlement must occur before the transcrystalline fracture changes to intercrystalline.

The question of how a difference between the two transition temperatures of a steel (*i.e.*, in the unembrittled and embrittled conditions) over one range of temperature is related to another difference over some other range of temperature has not previously been discussed quantitatively. To interpret the present results it is proposed to assume that the variation of yield stress (under the dynamic conditions of the impact test) with absolute temperature is linear, and that where a transition to intercrystalline brittle fracture occurs the relationship between the transition temperature T (° K.) and the intergranular cohesion C is:

$$C = a + \frac{b}{T}$$

On this basis, the decrease in intergranular cohesion

* The fact that the transition from tough to brittle fracture often extends over a range of temperature may be attributed to heterogeneous stress distribution and strain rate in the test piece, and does not affect the substance of the present hypothesis.

Table VI—FOURTH SERIES : EMBRITTLEMENT TEST RESULTS

Alloy	Composition	Unembrittled			Embrittled 670 hr. at 450° C.			Response to Zephiran Etchant
		Trans. Temp., °C.	V.H.N.	Brittle Fracture	Trans. Temp., °C.	V.H.N.	Brittle Fracture	
40	Fe–Cr–P–C	−101	232	Mainly trans-, some intercryst.	+ 6	223	All intercryst.	Good
45	Fe–Mn–P–C	−151	203	Mainly transcryst.	− 34	183	,, ,,	Nil
48	Fe–Mn–Cr–P–C	−112	253	Mainly intercryst.	+116	245	,, ,,	Good
49	Fe–Mn–Ni–P–C	−137	214	Mainly trans-, tr. intercryst.	+ 29	213	,, ,,	Nil
60	Fe–Cr–Ni–P–C	−122	229	Mixed inter- and transcryst.	+ 10	233	,, ,,	Good
68	Fe–Cr–P–C	− 76	217	Mainly inter-, tr. transcryst.	+ 33	217	,, ,,	Good

between the equilibrium states at 650° and 450° C. (now referred to as the modified susceptibility S) is given by:

$$S = b\left(\frac{1}{T_{650}} - \frac{1}{T_{450}}\right),$$

where T_{650} and T_{450} are the equilibrium transition temperatures (° K.) for transition to intercrystalline brittle fracture after treatment at 650° and 450° C. respectively (and in this case after 1 hr. at 650° C. and 670 hr. at 450° C.). To compare the susceptibilities of various alloys it will be necessary to make the further reasonable assumption that the constant b is the same for all the alloys investigated. Table VII summarizes the transition temperatures of the six susceptible alloys of this series in both the unembrittled state (t_{650}) and in the state of maximum embrittlement (t_{450}). The susceptibility of each alloy is expressed as the simple difference between the two transition temperatures, and is now referred to as the apparent susceptibility; the modified susceptibility S, as defined above, is also given.

The results in Table VII lead to the following conclusions:

Relative Embrittling Powers of Manganese and Chromium (Comparison of Alloys 40 *and* 45)—In the presence of a given concentration of phosphorus, the apparent and the modified susceptibilities are greater with Mn than with Cr, on the basis of either atomic or weight percentage. This is also true in the presence of 3% of Ni (comparison of alloys 49 and 60).

Effect of Manganese and Chromium Together (Comparison of Alloys 40, 45, *and* 48)—The alloy containing both Mn and Cr (48) shows the greatest value for apparent susceptibility, though not the greatest value for modified susceptibility. The available data do not enable very definite conclusions to be drawn, but the value of S is rather more than would be expected if either Mn or Cr were absent. It appears that both Mn and Cr contribute to the fall in intergranular cohesion on embrittlement, but not in a purely additive fashion, for this could be expected to give a value of S of *at least* 450 (250 due to 2·3% Cr and 200 due to 0·99% Mn).

Effect of Nickel in the Presence of Manganese (Comparison of Alloys 45 *and* 49)—The addition of 3% of Ni greatly increases the apparent susceptibility, but values of S indicate that Ni has no real effect upon the fall in intergranular cohesion. Thus, although Ni increases susceptibility from a practical point of view, its effect is probably not one of increasing the actual extent of the grain-boundary change.

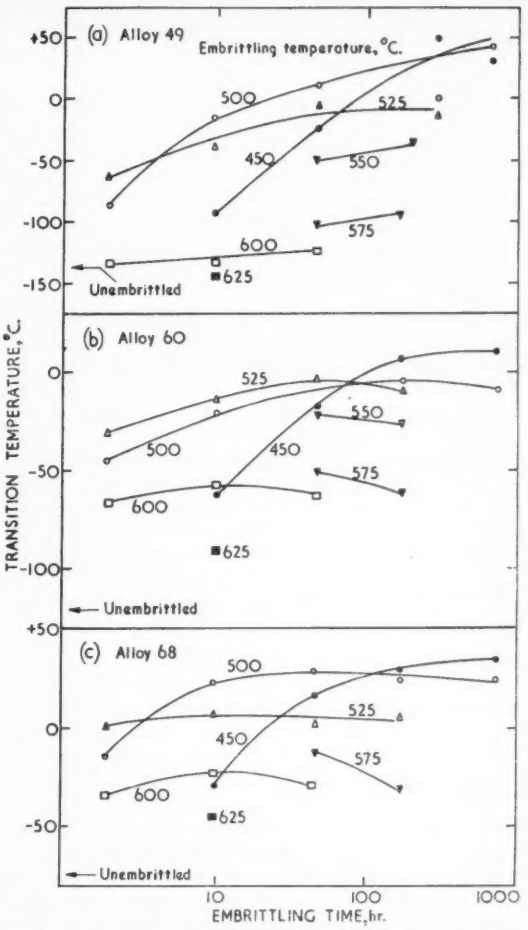


Fig. 20—Isothermal embrittling characteristics of alloys 49, 60, and 68 (series 4)

Effect of Nickel in the Presence of Chromium (Comparison of Alloys 40 *and* 60)—The addition of 3·2% of Ni slightly increases the apparent susceptibility.

Table VII

SUMMARY OF TRANSITION TEMPERATURES

Alloy	Composition	Transition Temperatures, ° C.			$S \times 10^5$
		t_{650}	t_{450}	$t_{450}-t_{650}$	
40	2·4% Cr, 0·044% P	−101	+ 6	107	223*
45	1·36% Mn, 0·048% P	−151	− 34	117	402***
48	0·99% Mn, 2·31% Cr, 0·053% P	−112	+116	228	364
49	1·6% Mn, 3·0% Ni, 0·045% P	−137	+ 40	177	416**
60	2·13% Cr, 3·2% Ni, 0·058% P	−122	+ 10	132	309
68	1·98% Cr, 0·094% P	− 76	+ 33	109	181

One, two, or three asterisks against values of S indicate that the brittle fracture in the unembrittled state of that alloy was not predominantly intercrystalline. In those cases a predominantly intercrystalline brittle fracture first occurred at the following embrittling temperatures: * 625° C.; ** 600° C.; *** 575° C. Thus the number of asterisks gives a qualitative indication of the amount by which the quoted value of S falls short of the true value.

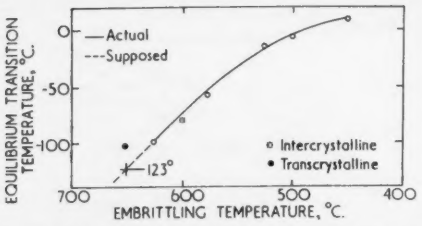


Fig. 21—Variation of equilibrium transition temperature with embrittling temperature for alloy 40

Comparison of values of S shows a slight increase in modified susceptibility due to the presence of Ni, but allowance must be made here for the fact that the quoted value of S for alloy 40 is low, since its brittle fracture in the unembrittled state is mainly transcrystalline. For alloy 40 to give the same value of S as alloy 60, its transition temperature for *intercrystalline* brittle fracture in the unembrittled state would have to be − 123° C. Extrapolation of the graph showing equilibrium transition temperature against embrittling temperature (Fig. 21) indicates that this is quite possible, and hence that Ni has no effect upon the actual fall in intergranular cohesion on embrittlement.

Over-ageing

All the alloys showed some decrease in hardness after lengthy treatments at 550° C. and above, and this could to some extent be the reason for the fall in transition temperature in some cases after long treatments above 525° C. This fall occurs almost exclusively in the alloys containing Cr, and not in the two Cr-free Mn alloys (45 and 49). It is unlikely that errors in temperature measurement or control are responsible for the changes, since alloys 40 and 48, which both show indications of over-ageing, and alloys 45 and 49, which do not, were all heat-treated together. The slower approach to equilibrium of the Mn alloys at higher embrittling temperatures may be connected with these differences between them and the Cr alloys. Since a fall in hardness may cause the decreases in transition temperatures concerned, the present experiments provide only slight evidence of over-ageing.

Etching Behaviour

It has been confirmed that temper-brittleness can be detected by picric acid etchants in Cr-containing alloys, but not in susceptible alloys containing Mn but no Cr; this is also true with 3% of Ni present. The typical dark-etching grain boundaries and subboundaries that were readily revealed in susceptible Cr alloys were not observed in susceptible Mn-containing alloys when Cr was absent.

INFLUENCE OF CARBON

Effect of Lowering the Carbon Content

The effect of lowering the carbon content of an Fe–Cr–P–C alloy from 0·21 to 0·08% is as follows:

Alloy 40 (2·4% *Cr*, 0·044% *P*)

	Unembrittled		Embrittled 48 hr. at 500° C.	
Carbon, %	V.H.N.	Trans. Temp., ° C.	V.H.N.	Trans. Temp., ° C.
0·21	232	−101	233	−16
0·08	182	− 85	183	−39

These results indicate that an increase in carbon is likely to favour the appearance of temper-brittleness under given conditions, though not necessarily that the actual fall in intergranular cohesion on embrittlement is less, for the change in mechanical properties of the bulk of the alloy resulting from the lower carbon content will be expected to cause a change in the transition temperature associated with any particular value of intergranular cohesion. The lower yield stress of the lower-carbon material would be most likely to cause such a change in a downward direction, which is in fact observed. This is consistent with (though not proving) the view that the fall in intergranular cohesion accompanying embrittlement is independent of carbon content within the range examined.

Reduction of Carbon Content to Near the Limit of Solid Solubility

Specimens of two Cr-containing alloys were carburized to approximately 0·014% of C, and were impact-tested in the unembrittled condition and after 48 hr. at 500° C. The compositions of the two alloys and the transition temperatures are given below:

Alloy	Composition	Unembrittled Trans. Temp., ° C.	Embrittled 48 hr. at 500° C. Trans. Temp., ° C.
19	1·6% Cr, 0·002% P, 0·014% C	Below − 110	Below − 110
16	1·4% Cr, 0·04% P, 0·014% C	,,	− 75

Alloy 19, which was non-susceptible with 0·1% of C, showed practically no dark-etching grain boundaries in either condition of heat-treatment; with the susceptible alloy 16, the zephiran etchant gave clear differentiation between the unembrittled and embrittled conditions, as shown in Figs. 11 and 12. The embrittling change was therefore still proceeding at this low carbon content.

Reduction of Carbon Content to Below the Limit of Solid Solubility

Specimens of the first series of alloys were treated for 90 hr. in an atmosphere of hydrogen containing 25% by volume of water vapour, after the method of Low and Gensamer.[5] The resulting carbon and nitrogen contents were less than 0·001%. After the usual treatments to give the unembrittled and embrittled conditions, the zephiran reagent gave some occasional indications of the embrittling change in the Fe–Cr–P alloy, but transition-temperature determinations gave no consistent indication of temper-brittleness. A predominantly intercrystalline brittle fracture was found in almost all the alloys in all conditions of heat-treatment.

At this stage it became evident that very small concentrations of oxygen might cause intercrystalline weakness in carbon-free alloys. The two possible sources of contamination with oxygen, both of which could be responsible for the introduction into carbon-free alloys of sufficient oxygen to cause an intercrystalline brittle fracture, were (i) the moist hydrogen atmosphere used for decarburization, and (ii) very small residual concentrations of air in the furnace used for vacuum heat-treatment. Only treatments above 650° C. (especially at 1000° C.) in the vacuum heat-treatment furnace gave serious intercrystalline

brittle fracture, and the initial water-quenching from 1000° C. was eliminated in future heat-treatments since it was not essential for the development of temper-brittleness in susceptible materials. Because of the danger of introducing oxygen with a moist hydrogen atmosphere, a dry decarburizing atmosphere was used. In early experiments the carbon content of pure iron at 950° C. in equilibrium with dry commercial hydrogen was about 0·01%, and since the carbon content of iron in equilibrium with a given carburizing atmosphere should be considerably lower for α- than for γ-iron, it seemed probable that treatment for a week in dry hydrogen at 950° C. (this removed oxygen sufficiently in many cases to eliminate the persistent intercrystalline brittle fracture due to oxygen), followed by a week in the same atmosphere at 750° C., would give a material with very low carbon and oxygen contents.

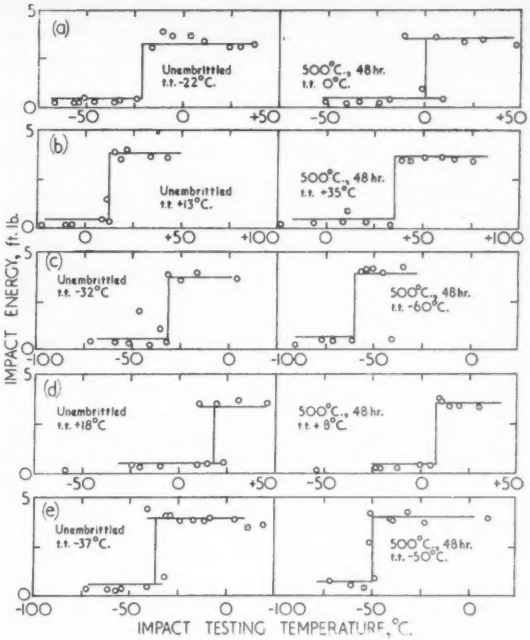
Specimens of five alloys were treated in this way and their carbon contents (using 50-g. samples and a preliminary acid copper potassium chloride separation) were estimated at approximately 0·003%. The alloys were then heat-treated as follows, in an attempt to introduce temper-brittleness:

Unembrittled state: 650° C. (1 hr.), W.Q.
Embrittled state: As above, then 48 hr. at 500° C., W.Q.

Care was taken to attain a good vacuum in the heat-treatment furnace before the specimens were raised into the hot zone, and the slight tarnish produced by quenching into water from 650° C. was removed before the embrittling treatment at 500° C. The transition temperatures of these materials and the nature of their brittle fractures are given in Table VIII. The variation in impact energy with testing temperatures is shown in Fig. 22.

Where scatter in the impact tests occurred, the transition temperature has been taken as halfway between the highest temperature of brittle fracture and the lowest temperature of tough fracture. There was no significant difference in the response of any of the alloys to the zephiran etchant between the unembrittled and embrittled conditions. Alloy 40 (Fe–Cr–P) developed very faint, dark grain boundaries in both states, but they were of equal intensity in both, whether etched lightly or deeply, while darker boundaries were visible near the edges of the specimens, suggesting that a trace of oxygen had entered.

If the embrittling change occurs in susceptible alloys



(a) Alloy 44 (d) Alloy 40
(b) Alloy 14 (e) Alloy 42
(c) Alloy 22

Fig. 22—Impact-energy diagrams for low-carbon alloys

in the absence of carbon, and if its occurrence under such conditions lowers the intergranular cohesion in the same way as in carbon-containing alloys, then where an intercrystalline brittle fracture occurred in the unembrittled state it would be expected that temper-embrittlement would cause a rise in the transition temperature. The results (Table VIII) give no consistent indication of such a change. Although the susceptible Fe–Mn–P alloy (14) showed a slight rise in transition temperature, alloy 40, which in the presence of 0·21% of carbon gave a rise of 85°, showed no rise at all in the absence of carbon. The etching test, which in the presence of carbon gives a very consistent and reliable indication of temper-brittleness, would be expected to respond similarly in the absence of carbon if the embrittling change still occurred. Such an indication was never observed

Table VIII

LOW-CARBON ALLOYS: EMBRITTLEMENT TEST RESULTS

Alloy	Composition	Unembrittled		Embrittled		Difference, $t_e - t_u$, °C.
		Trans. Temp. (t_u), °C.	Brittle Fracture	Trans. Temp. (t_e), °C.	Brittle Fracture	
44	1·98% Mn	−22	Mainly intercrystalline	0	All intercrystalline	+22
14	1·4% Mn, 0·01% P	+13	All intercrystalline	+35	,, ,,	+22
22	1·47% Cr	−32	Mainly cleavage	−60	,, ,,	−28
40	2·4% Cr, 0·044% P	+18	Mainly intercrystalline	+ 8	Mainly intercrystalline	−10
42	2·4% Cr, 0·045% P, 0·51% Mo	−37	All cleavage	−50	All cleavage	−13

in the carbon-free alloys where precautions had been taken to obtain a low oxygen content, and this is regarded as good evidence that the embrittling change does not occur under those conditions.

Examination of the brittle fractures of these low-carbon, low-oxygen materials also gave no satisfactory indication of temper-brittleness in the absence of carbon. The mainly cleavage brittle fractures of the normally (*i.e.*, in the presence of carbon) non-susceptible Fe–Cr and Fe–Cr–Mo–P alloys (22 and 42) in comparison with the mainly intercrystalline brittle fracture of the Fe–Cr–P alloy (40), which with carbon present would be susceptible, might indicate the embrittling change in the latter in the absence of carbon; but in view of the complete lack of rise in transition temperature and response to the zephiran etchant, this evidence cannot be regarded as very significant.

In a further experiment, three normally susceptible Cr-containing alloys, heat-treated in hydrogen to avoid the introduction of oxygen, failed to show any increase in transition temperature on embrittlement, and there was no differentiation between the unembrittled and embrittled conditions on etching with the zephiran reagent.

It was concluded that the embrittling change does not occur in the absence of carbon, and that the concentration of carbon required for its occurrence appears to be of the order of the solid solubility limit of carbon in ferrite.

ELECTRON MICROSCOPE EXAMINATION

Susceptible Cr-containing alloys showed a fine dark-etching region at grain boundaries and sub-boundaries when etched with picric acid in either aqueous or alcoholic solution. The zephiran reagent accelerated this action. The development of this change on treatment at 500° C. for 110 hr. is shown in Figs. 13 and 14. Figures 15 and 16 are electron micrographs of the same specimens, which show the affected grain-boundary region etched into a diffuse channel, with no resolution of particles. The width of this channel in some of the sub-boundaries is no more than 100 Å. The dark channel is quite distinct from and apparently unrelated to the carbide particles.

LOCATION OF THE EMBRITTLING CHANGE

A miniature impact specimen of the carburized alloy 23 (Fe–Cr–P–C), which showed distinct susceptibility to temper-brittleness and a clear response to the zephiran etchant, was severely cold-rolled in the unembrittled condition. It was then heated for 1 hr. at 650° C. and water-quenched, to cause recrystallization and also to keep the material in the unembrittled state. Half this specimen was reheated to 500° C. for 48 hr. Micro-examination, using the zephiran etchant, showed that the treatment at 650° C. had caused partial recrystallization, and that the zephiran etchant revealed a dark-etching network, characteristic of temper-brittleness, only in the half that had been heated at 500° C. (Figs. 17 and 18). The occurrence of the embrittling change is shown in Fig. 18 equally at the old distorted grain boundaries, which represent the location of the prior austenitic grains, and at the new grain boundaries formed by recrystallization and therefore unrelated to any prior austenitic structure. This shows clearly that the temper-embrittling change occurs at ferritic grain boundaries, whether they represent the prior austenitic structure or not.

GENERAL DISCUSSION AND CONCLUSIONS

In considering the nature of the embrittling process the following conclusions must be taken into account:

(1) Temper-brittleness does not occur in Fe–C or Fe–C–P alloys.

(2) The addition of Mn or Cr to an Fe–C–P alloy causes the embrittling change to occur. The minimum concentration of these elements required to induce temper-brittleness is not known, but the embrittling change can be detected by etch tests in alloys containing only 0·4% of Cr + 0·01% of phosphorus.

(3) With about 2–2·5% of either Mn or Cr, the embrittling change can occur to a slight degree with $< 0{\cdot}001\%$ of phosphorus.

(4) An increase of phosphorus in the Fe–Mn–C and Fe–Cr–C alloys, at least in the lower ranges (say, up to 0·04%), increases the degree of the embrittling change, as does also an increase in Mn or Cr.

(5) Removal of carbon from susceptible alloys to at least 0·014% has little effect upon embrittlement, though transition temperatures are indirectly affected, but this is due to a change in the mechanical properties of the steel matrix rather than to a change in the embrittling reaction. Reduction of the carbon to 0·003%, however, eliminates the embrittling change.

(6) The presence of Mo greatly reduces the equilibrium extent of the change at a given embrittling temperature, but does not appear to influence the rate of embrittlement.

(7) The process of temper-embrittlement involves a change at the ferritic grain boundaries and sub-boundaries of susceptible materials. The location of the embrittling change is independent of any prior γ-iron (austenitic) structure and proceeds only at the grain boundaries and other lattice discontinuities. From micrographic evidence, the region involved in the change would seem to occupy a thickness of no more than 1000 Å. and quite possibly less than 100 Å., and suffers a decrease in cohesive strength.

(8) There is no marked tendency towards over-ageing, and micrographic evidence (in the present work and elsewhere[6]) has not shown any tendency for the grain-boundary constituent to agglomerate during very long embrittling treatments.

Any theory which attempts to explain temper-brittleness on the basis of a simple precipitation from supersaturated solid solution is unlikely to be satisfactory, particularly in view of the lack of changes in mechanical properties normally associated with precipitation processes. The reversibility of the process, and the persistence of the change at grain boundaries and other lattice discontinuities, strongly suggests that the process is essentially one that can occur and persist only at grain interfaces. Two possible processes are:

(1) A segregation of some element(s) to the grain boundaries, leading to the building-up of a higher concentration of the segregating constituents at the grain boundaries than within the grains, but without

the separation of a new phase; (2) the formation of a new phase at the grain boundaries and other lattice discontinuities, existing there in a state of equilibrium, possibly as suggested by Jaffe.[7]

The etching difference between Mn and Cr alloys is important, for the embrittling process clearly occurs with almost identical characteristics in both alloys, yet picric acid etchants differentially attack the changed regions in the Cr-containing alloys only. If a process of equilibrium segregation is involved, as McLean and Northcott[6] suggest, the present findings would indicate carbon as the element involved, and if this were the case the etching differences would be very difficult to explain. Furthermore, the picric acid etchants detect the embrittling change in Cr-containing alloys even at very slight degrees of embrittlement, and fail to do so in Mn-containing alloys even after severe embrittlement. This fact might be adduced in support of the theory of the separation of a new phase, rather than to one of equilibrium segregation in solid solution, for in the former case the electrode potential of the embrittled region will be independent of the degree of embrittlement, while in the latter it will vary with the concentration of the embrittling constituent.

Although the present findings give some support to the belief that the separation of some definite constituent occurs at the grain boundaries, they do not enable such a phase to be definitely identified. It appears that the presence of carbon, together with Mn or Cr, is essential for susceptibility. Although there is a considerable amount of support for the belief that the presence of phosphorus (or As) is essential, the occurrence of temper-brittleness in the almost phosphorus-free alloys, particularly in the Fe–Mn–C alloy (44), suggests that the change may occur in the absence of phosphorus, although phosphorus is nevertheless a powerful susceptibility-inducing element.

A reasonable hypothesis that can be tentatively proposed from the present results is that the embrittling process involves the formation of an extremely fine dispersion or film of a carbide of Mn or Cr at the ferritic grain boundaries and sub-boundaries. The equilibrium concentration of these carbides is increased by increasing the Cr, Mn, P, As, or carbon concentration within the grains and also by decreasing the embrittling temperature (although below about 450° C. the time required for equilibrium to be approached is extremely long). The influence of the strongly carbide-forming element Mo in inhibiting the embrittling change may be attributed to the formation of relatively more stable carbides within the grains. Nickel has not been found to induce susceptibility in alloys free from Mn and Cr, and this is in accord with its lower carbide-forming tendency.

It is of some interest to consider what is the rate-limiting process in the development of temper-brittleness. Since the embrittling change occurs to some extent even at the pre-embrittling tempering temperature of 650° C., the process is unlikely to be retarded at lower temperatures for want of nuclei, and the rate-controlling process must be diffusion of some element in ferrite. Estimates of the activation energy of the process have placed it at a rather higher value than that for the diffusion of carbon in α-iron (46,000 cal. for temper-embrittlement, Jaffe and Buffum[8]; 18,000 cal. for carbon in α-iron, Stanley[9]), showing that the rate of diffusion of some element other than carbon in ferrite governs the rate of development of temper-brittleness, and suggesting that the diffusion of Mn or Cr is the rate-controlling process. Also, the high value of the activation energy for temper-embrittlement, together with the fact that the presence of carbon is essential for its development, is in accord with the lack of susceptibility of Fe–C alloys, in which the rate-limiting process could only be that of diffusion of carbon in α-iron.

SUMMARY

(1) Temper-embrittlement involves a change at the ferritic grain boundaries (and sub-boundaries) of susceptible materials. The location of the embrittling change is independent of any prior γ-iron (austenitic) structure. The region involved in the change appears to be less than 1000 Å. in thickness, and possibly less than 100 Å., and it suffers a decrease in cohesive strength. The grain-boundary change may lead to an intercrystalline brittle fracture and a corresponding rise in transition temperature in the impact test, although the embrittling change can proceed slightly before the fall in intergranular cohesion is sufficient for an intercrystalline brittle fracture.

(2) Temper-brittleness has not been detected in a pure Fe–C or Fe–P–C alloy, but it occurs to a marked extent in iron-base alloys containing Cr or Mn, together with phosphorus and carbon. It also occurs slightly in almost phosphorus-free ($< 0{\cdot}001\%$) alloys containing high Mn or high Cr and carbon.

(3) Temper-brittleness may be detected by the etch test in Fe–Cr–P–C alloys containing as little as $0{\cdot}4\%$ of Cr and $0{\cdot}014\%$ of phosphorus, but in such alloys (containing no Mn) about $0{\cdot}8\%$ of Cr has to be present before the embrittling change leads to intercrystalline brittle fracture and a rise in transition temperature.

(4) On the basis of either weight or atomic percentage, the embrittling effect of Mn is appreciably greater than that of Cr.

(5) There is little difference in the kinetics of the embrittling process between susceptible Mn-containing and Cr-containing alloys, apart from a somewhat more rapid approach to equilibrium at higher embrittling temperatures with the latter.

(6) Temper-brittleness was not detected in phosphorus-free Fe–Ni–C alloys containing up to 4% of Ni, nor in an iron–carbon ($0{\cdot}1\%$ C) alloy containing $0{\cdot}04\%$ of phosphorus and $1{\cdot}5\%$ of Ni.

(7) The addition of Ni to susceptible Fe–Mn–P–C and Fe–Cr–P–C alloys does not appear to affect the fall in intergranular cohesion on embrittlement, though it may make temper-brittleness more apparent in the impact test.

(8) The addition of about $0{\cdot}5\%$ of Mo to susceptible Mn- or Cr-containing alloys almost completely suppresses the embrittling change, its effect being to greatly decrease the equilibrium extent of embrittlement, rather than the rate at which equilibrium is attained.

(9) Reducing the carbon content of a susceptible

alloy at first has little effect upon the development of the embrittling change, although the altered mechanical properties of the base material tend to make the alloy appear less susceptible in the impact test; with carbon reduced to about 0·003%, susceptibility disappears.

(10) The presence of As in a phosphorus-free Fe–Cr–C alloy produces marked susceptibility.

(11) The development of the embrittling change may be detected in susceptible Cr-containing alloys by aqueous or alcoholic picric acid etchants, and the addition of surface-active reagents increases the sensitivity of such etchants. Although the embrittling change occurs with almost identical characteristics in Mn-containing alloys, it cannot be detected by these same etchants in the absence of Cr.

Acknowledgments

Grateful acknowledgment is made to the British Iron and Steel Research Association for financial support during the investigation; to Mr. R. S. Brown for arranging for the alloys to be drawn down to wire, and also to the British Non-Ferrous Metals Research Association for the electron micrographs included in this report.

References

1. R. F. Wells and C. Mehl: *Trans. Amer. Inst. Min. Met. Eng.*, 1937, vol. 125, pp. 429–469.
2. J. B. Cohen, A. Hurlich, and M. Jacobson: *Trans. Amer. Soc. Metals*, 1947, vol. 39, pp. 109–136.
3. J. H. Hollomon: *Ibid.*, 1946, vol. 36, pp. 473–540.
4. S. A. Herres and A. R. Elsea: *J. Met.*, 1949, vol. 1, No. 6, pp. 366–370.
5. J. R. Low and M. Gensamer: *Trans. Amer. Inst. Min. Met. Eng.*, 1944, vol. 158, pp. 207–242.
6. D. McLean and L. Northcott: *J. Iron Steel Inst.*, 1948, vol. 158, pp. 169–177.
7. L. D. Jaffe: *Ibid.*, 1950, vol. 164, pp. 1–3.
8. L. D. Jaffe and D. C. Buffum: *Trans. Amer. Inst. Min. Met. Eng.*, 1950, vol. 42, pp. 604–614.
9. J. K. Stanley: *J. Met.*, 1949, vol. 1, No. 10, pp. 752–761.

ANDREW CARNEGIE SCHOLARSHIP REPORT

The Deformation of Austenite in Relation to the Hardness Characteristics of Steel

By G. R. Bish, M.Sc., and H. O'Neill, D.Sc., M.Met.

SYNOPSIS

The pyramid hardness of a 0·9% carbon steel during isothermal transformation at two temperatures just above the M_s line has been determined. The same steel was plastically deformed by crushing during the incubation period of isothermal transformation at these two temperatures. For various amounts of deformation and increasing holding times in the isothermal bath, the hardness at room temperature of the quenched products has been determined by Meyer analysis. Deformation accelerates transformation, and the Meyer index n passes through a maximum with increasing holding times. The maximum occurs at shorter holding times as the amount of deformation is increased. The hardness and strain-hardening properties of various products of heat treatment have been reviewed, and a complete hardness diagram for the Fe–Fe_3C system is given in an Appendix. 737

THE deformation of steel while it is undergoing the austenitic transformation is a matter of practical and theoretical importance. Whether or not it is true that Biringuccio[1] appreciated in 1540 the improved product that resulted from interrupted quenching treatments, there is the practical suggestion of another pioneer in this field that should be remembered. Lewis[2] pointed out in 1929 that, with a single heating, steel might be shaped or coiled during transformation; it could then cool down to finish in a very hard condition. The effect of plastic strain on the transformation has been considered to be worthy of further study, because optimum properties might result from certain combined treatments of the austenite. It is known, for instance, that some austenitic steels undergo certain changes when cold-worked. Progressive rolling of an 18/8 alloy at room temperatures first reduces the Meyer index n of the ball indentation test in the usual way, as shown in Table I,[3] but there is eventually a slight rise. Post and Eberly[4] have reported that reducing an austenitic steel 80% produces pseudo-martensite, which may be associated with the inversion just mentioned. Jones[5] has shown that the rate of straining affects the work-hardening of 18/8 alloy, and that a value of n of 2·38 in the condition as-softened from 1100° C. (153 Brinell) falls to 2·22 as carbides are precipitated by reheating at 500° C. (262 Brinell). The fall in the value of n on reheating a high-carbon austenitic steel may also be observed in Table I.

Hadfield manganese steel presents some problems;

Manuscript first received 2nd September, 1952, and in its final form on 16th January, 1953.

Mr. Bish is metallurgist to Enfield Cables, Ltd., and Dr. O'Neill is Professor of Metallurgy at University College, Swansea.

Table I

CHANGES IN AUSTENITIC STEELS

Steel	Treatment	Meyer Index n	Ultimate Meyer Hardness P_u, kg./sq. mm.	Brinell Hardness Number ($L/D^2 = 30$)
18% Cr, 8% Ni	No cold-reduction	2·51	261	169
	10% " "	2·19	354	268
	19% " "	2·18	380	298
	60% " "	2·24	625	444
1·9% C, 2·0% Mn	W.Q., 1100° C.	2·43	370	241
	W.Q., tempered at 400° C.	2·38	854	495
	W.Q., tempered at 900° C.	2·24	610	415
	Normalized	2·26	541	396

Goss[6] found that 87% cold-reduction caused a rise in hardness to *C*58 Rockwell (about 650 V.D.H.), although the alloy was still austenitic when examined by X-ray diffraction.

Some effects of deformation on carbon-steel austenites during isothermal transformation have been published[14] since the present work was begun, but none of the investigators has used the method of Meyer hardness analysis, based on the expression $L = ad^n$. This procedure gives more information than the simple indentation test, and has previously[7] enabled the age-hardening of duralumin to be represented as a strain-hardening process. Furthermore, if volume changes during transformation produce strain-hardening effects, these might be revealed by a fall in the n index. Finally, by using the ultimate hardness $P_u(= 4a/\pi.D^{n-2})$, where D is the ball diameter, truly comparable values are obtained, which are almost independent of grain-size variations.

An outline of the hardness diagram for the iron–iron-carbide system is given in the Appendix; it indicates the possibilities of strain-hardening effects in hard alloys. The high values for hardened steels require the Meyer analysis to be conducted with a diamond ball, and a special 1-mm. hemispherical indenter was prepared for this work.*

MEYER ANALYSIS OF HARD STEELS

Previous experiments[3, 8] by one of the authors had shown that the Meyer index n increased appreciably when a pearlitic steel was converted to martensite. Subsequent micro-indentation tests by Rostoker,[9] however, using a diamond pyramid, showed not only an unexpected change of logarithmic index for such a constant-angle indenter, but also one in a sense contrary to that just mentioned for the ball test. The earlier macro-hardness work has therefore been repeated on another eutectoid steel *TX* of 1 in. dia. × $1\frac{1}{8}$ in. thick, using the diamond pyramid, and also a 1-mm. diamond hemisphere under concentric loadings of 20–50 kg. on the Vickers machine. Although this highest load sometimes produced fine radial cracks around impressions in martensite, subsequent structures containing bainite were generally free from cracking. It is unfortunate that, if cracking is to be

* Constructed by Mr. P. Whitaker, F.I.M.

Table II

HARDNESS VALUES FOR DIRECT AND INTERRUPTED QUENCHING

Treatment	1-mm. Diamond Ball		136° Diamond Pyramid	
	Meyer Index n	Ultimate Hardness P_u, kg./sq. mm.	Pyramid Hardness (30 kg.), kg./sq. mm.	Logarithmic Index 10–40 kg.
Steel *TX* (C, 0·93%; Mn, 0·63%; Si, 0·25%; Ni, 0·38%; Cr, trace)				
Normalized	2·31	445	323	...
W.Q. from 850° C.	2·41	1403	836	2·00
Quenched from 850° to 230° C., held for 300 sec., and W.Q.	2·51	1480	812	...
W.Q. from 850° C., tempered at 250° C. for 1 hr.	2·31	1048	...	2·00
W.Q. from 850° C., tempered at 450° C. for 1 hr.	2·16	538	...	2·00
Steel *TN* (C, 0·51%; Cr, 1·21%; Mn, 0·64%; Si, trace)				
W.Q. from 900° C.	2·41	865	714	...
Quenched from 900° to 230° C., held for 300 sec., and W.Q.	2·49	772	600	...

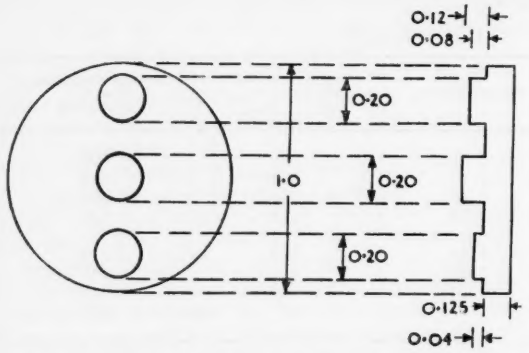


Fig. 1—Details of test disc (all dimensions in inches)

avoided, the upper limit of test load must be about 50 kg., because, although the indents reported here exceed the $d/D = 0{\cdot}1$ minimum for which Meyer reported his law to be valid, Tabor[10] has deduced that a minimum load of about 52 kg. would be necessary to bring hard steels into the range of fully representative plastic indentation. It will be appreciated that it is difficult to measure the small shallow indentations obtained with low loads, and so plotted values for hard steels may show some scatter.

The results are given in Table II, including some obtained for comparing direct and interrupted quenching. Steel *TX* had an inherent A.S.T.M. grain size by the McQuaid–Ehn test of 5. The macro-logarithmic index is 2·00, as would be expected for a pyramid, but for a similar eutectoid carbon steel given the same treatments, Rostoker obtained micro-index values (25–200-g. loads) of 3·34, 2·85, and 2·85 respectively, results that are difficult to explain. Subsequent work[11] with ten different micro-testers has shown that the micro-pyramid index approximates to 2·0 in both strain-hardened and quench-hardened metals.

In confirmation of earlier work, it is evident that the n value of 2·41 (20–50-kg. load) is higher for martensite than for the softer ferrite + carbide structures, and it is also higher for interrupted quenching (2·51) than for direct quenching. This would be consistent with the view that direct-quenched martensite is relatively in a more strain-hardened condition than that which has begun to transform isothermally. A bath temperature of 230° C. is above the martensite (M_s) line for steel *TX*, but below it for steel *TN*.

Rostoker doubted the work-hardenability of martensite, but Fig. 7 (*see* Appendix) shows that a quenched 0·9% carbon steel has been superhardened 40% by cold-working. Although this is less than the increase of 90% for pearlite, it is doubtful whether the martensitic specimen has been deformed to an equivalent extent.

APPARATUS FOR DEFORMATION

When this work was begun in 1948, no publication had been seen which dealt with the effect of strain upon austenite transformation in plain carbon steels. It was decided to compress special disc specimens of less than critical thickness by drop hammer at constant high speed while they were in the isothermal bath. Each disc specimen was prepared with three initial projections of different heights; these were intended to be forged to practically uniform final thickness by a single blow of the drop hammer, so as to produce three patches of steel that had been crushed by different amounts and also an undeformed edge portion. The final shape of test disc was chosen after much experimenting (*see* Fig. 1), and the areas of uniform hardness in each deformed patch were explored by extensive preliminary work on test discs of Armco iron. The same cast of Armco iron had been used by Balicki[12] for cold-drawing experiments, and from his hardness results it was possible to produce calibration curves equating measured crushing of the disc projections with equivalent percentage reductions in area.

The drop-hammer, with the integral austenitizing furnace (40), the isothermal bath (16) with stirrer, and the pyrometer (24), is shown in two views at right-angles in Figs. 2*a* and *b*. The specimen was held in the cradle (32) in an atmosphere of dry purified nitrogen in the inclined austenitizing furnace (40) with water-cooled ends; it was then tipped down the stainless-steel tube (26) on to the anvil (14) in the isothermal bath (16) by manipulating the enclosed rod (34). The tube was moved upwards to clear the hot hammer bar (23), which was then lowered on to the specimen, and, after about 15 sec., the weight (11) was released to strike the hammer bar and crush the test disc (15). After a definite holding time, the disc was rapidly withdrawn with tongs, quenched in water, and tested for hardness at room temperature.

Provision was also made for static hardness tests on plain discs at elevated temperature during transformation without any previous deformation. These revealed the beginning and the end of transformation; it was also intended to determine Meyer n values of austenites before transformation. The apparatus for static tests fitted on to the bath frame; it consisted (*see* Fig. 3) of a 12-in. vertical spindle (9), carrying a Vickers diamond (8) at its lower end, which was immersed in the isothermal bath. The indenter with its applied load (12) could be lowered on to the specimen by means of a lever (14) and cam (17), and, after unloading, the platform system could be traversed in both directions over the specimen by means of screws. A line of spaced indents was thus made at different intervals of time, and their sizes were subsequently measured after removal from the bath.

Many experimental difficulties were encountered during the work, not the least of which was the liability of the disc specimens to crack, owing to attack by the hot salt baths. Various quenching liquids, including silicones, were tried, but eventually a nitrate–nitrite mixture was used; finally, the specimens were usually nickel-plated for protection. These difficulties reduced the number of results obtained during the period of the Carnegie Scholarship.

HOT HARDNESS TESTS

These tests were not absolute; with a calibrated load of 20 kg. operating for 6 sec., the indentations were made through the flash nickel coating of the

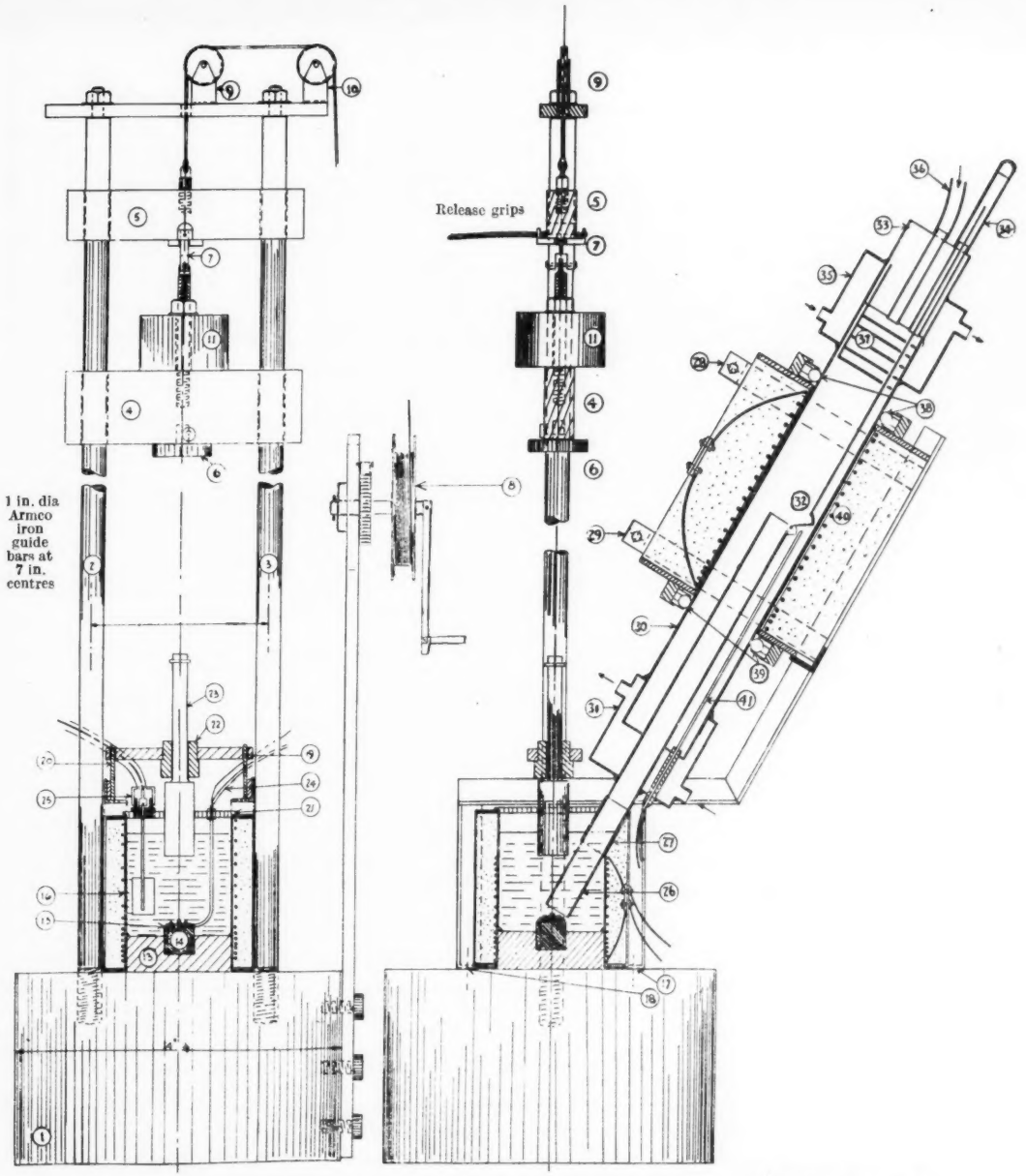


Fig. 2—Views at right-angles of drop-hammer, austenitizing furnace, and isothermal bath

plain disc specimens. Tests at room temperature on heat-treated steel *TX* (about 600 V.D.H.) showed that plating gave results that averaged 0·9 of those on the uncoated specimens. The hot hardness values shown in Fig. 4 have therefore been roughly corrected by this factor, and are perhaps slightly lower than standard.

Two isothermal temperatures of 280° and 245° C. were chosen at which to test the 0·9% carbon steel, because the M_s temperature should be about 190° C., and since hard final products were desired. Two sets of determinations were made at each temperature, and the transformation started after 180 and 360 sec. respectively. It was completed after about 1710 sec. at 280° C., but, although continued for 3500 sec. at 245° C., it is not certain whether 100% bainite was reached exactly at this time. There was a slight fall in the hardness values at the end of transformation, but it is not known whether this is significant. Metallographic examination confirmed the beginning of the precipitation, and the slight rise of hardness during the period of induction is noteworthy.

Enos, Peer, and Holtzwarth[13] have reported hardness tests made with a spring-loaded hardened-steel

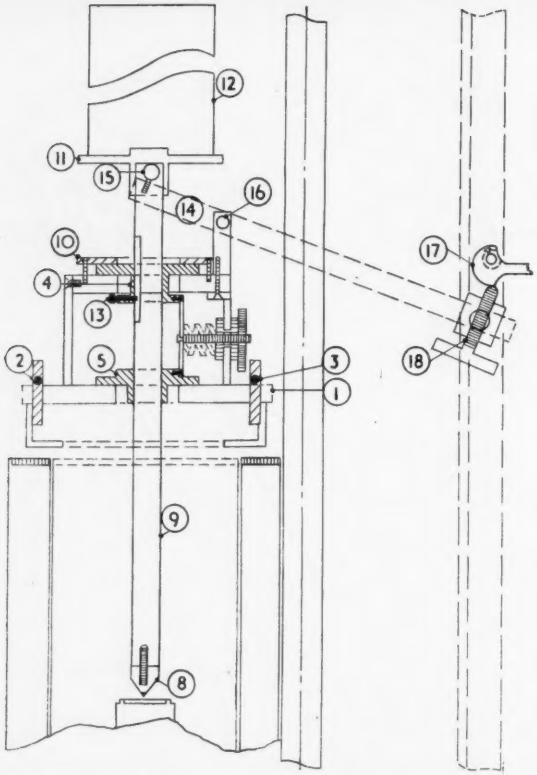


Fig. 3—Apparatus for determining hardness values of austenite during transformation

cone on an alloy die steel that was transforming isothermally; they found that the beginning and end of the reaction occurred at times agreeing with those

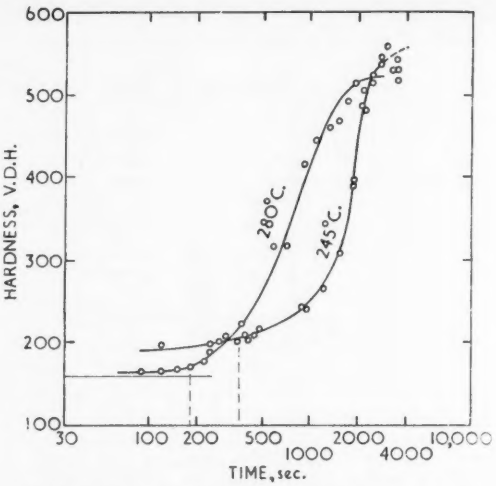


Fig. 4—Results of hot hardness tests on steel *TX* during transformation

determined by dilatometry and micrography. Their results show a constant hardness immediately after the completion of transformation.

EFFECT OF DEFORMATION DURING TRANSFORMATION

After 30 min. at 850° C., nickel-plated special disc specimens were crushed under the 9·4-kg. hammer falling 177 cm. at 280° C. in some cases and 245° C. in others (after 35 and 65 sec. delay respectively). They were allowed to transform in the bath for various holding times before being withdrawn, water-quenched, and indented. The deformation was applied at what were considered to be about the same fractional times in the induction period for the two

Table III

DEFORMATION AT 280° AND 245° C. WITH QUENCHING AFTER VARIOUS HOLDING PERIODS

Holding Time, sec.	0% Deformation		13% Deformation		41% Deformation	
	Pyramid Hardness, kg./sq. mm.	Bainite, %	Pyramid Hardness, kg./sq. mm.	Bainite, %	Pyramid Hardness, kg./sq. mm.	Bainite. %
Held at 280° C.						
145	812	...	886	5	840	20
175	801	trace	802	10	785	30
305	811	10	785	15	759	50
400	804	15	...	...	697	70
490	807	20	715	25–30	715 (?)	85
970	750	60	685	80	685	100
1325	713	75	645	100	675	100
1750	652	90–100	620	100	675	100
Held at 245° C.						
210	814	0	825	3	854	10
360	806	trace	822	7	836	15
900	820	15	800	25	787	60
1800	798	30	780	60	728	80
2950	760	50 (?)	752	80	698	100

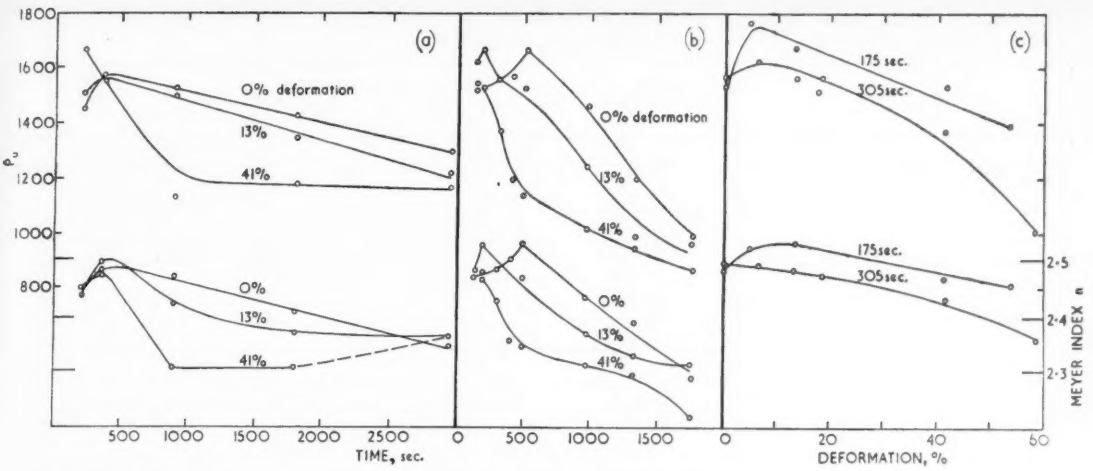


Fig. 5—Hardness values at room temperature after various deformations and various holding times before quenching. Temperature of deformation (*a*) 245° C., (*b*) and (*c*) 280° C.

bath temperatures. At the impact energies used, it was found that the smallest projection on each disc did not become deformed, and so in further tests drops of 224 and 230 cm. were used. After the nickel plating had been removed by polishing, the various areas were examined microscopically, and Meyer analysis was made at room temperature on appropriate parts. The results calculated from logarithmic plots are given in Figs. 5*a–c*.

The hardness of the final quenched products generally passes through a maximum, which usually occurs at shorter holding times as the amount of applied deformation increases. There is apparently a degree of transformation that gives a product which has maximum hardness and n values. This combination is a very unusual phenomenon in the treatment of alloys, and a warning must be issued about the difficulty of accurate n determinations on these hard steels. According to Fig. 5*c*, holding for 175 sec. after 5% deformation at 280° C., followed by quenching, gave optimum results from a hardening point of view, and the P_u value of 1750 kg./sq. mm. is greater than that given for direct quenching in Table II.

Some Vickers pyramid values are included in Table III which tend to fall continuously with holding time rather than to show the maxima provided by the P_u values. This is probably due to the geometrical features of the ball test.

MICROSTRUCTURES

The amount of transformation product under various conditions was observed by metallographic examination of the specimens, but no exact quantitative determinations were made. The percentage amounts of transformation product are therefore reported in Table III with some diffidence.

Many microstructures were photographed, and those shown in Figs. 6*a–f* have been selected for reproduction. Where small deformations have been applied, the bainite is located on what appear to be slip planes; this has also been observed by Jepson and Thompson.[14] Thus, Fig. 6*d* gives an effect resembling multiple twinning, compared with the indiscriminate chevrons of the undeformed specimens (Fig. 6*a*). With increased deformation (and transformation) the major slip planes seemed to be the origins of families of needles (Figs. 6*e* and *f*), but the slip-plane structures were not evident when transformation was practically completed (Fig. 6*c*).

DISCUSSION AND SUMMARY

Meyer analysis shows that, within the range of these experiments, the highest value of ultimate hardness was not obtained by water-quenching either normal or supercooled austenite before transformation starts. Better results follow if isothermal transformation is first allowed to produce about 10% bainite. The highest hardness was obtained when supercooled austenite was plastically deformed by compression just before partial isothermal transformation and quenching. These maximum hardness values were associated with superior n values, as if the steel finally remained in a relatively low condition of strain-hardening. The actual cold deformation applied would itself normally produce a slight fall in the n value, and a very slight increase in volume. Since the transformation of austenite involves an increase in volume, it is suggested that the plastic deformation energizes the transformation, but also has what might be called a compensating relaxation effect.

The maxima in the Meyer properties occurred at significantly differing and decreasing amounts of 'cold work'. This occurred at both temperatures chosen for isothermal transformation, *i.e.*, 280° and 245° C., but to a lesser degree at the lower temperature. Since the deformation would presumably create sites for nucleation of the bainite phase, increasing strain would be expected to make more sites available. That these maxima do occur at decreasing percentages of bainite would therefore appear to indicate that the applied strain is acting in

Table IV
TYPICAL VALUES

Treatment	Ultimate Hardness P_u, kg./sq. mm.	Meyer Index n	Diamond Pyramid Hardness, kg./sq. mm.	Bainite, %	Terms
Quenched to 280° C., compressed 5%, held for 175 sec., W.Q.	1750	2·53	(886)	5	Flow marquenching
Quenched to 280° C., held for 300 sec., and W.Q.	1480	2·51	812	0	Marquenching
W.Q. from 850° C.	1403	2·41	836	0	Direct quenching
Quenched to 280° C., held for 1750 sec., and W.Q.	984	2·29	652	100	Ausquenching
W.Q. from 850° C., tempered at 450° C. for 1 hr.	538	2·16	...	0	Tempering

a manner that tends to relieve the highly stressed austenitic phase.

The most significant results have been collected in Table IV, together with terms for referring to typical operations.

The work indicates that supercooled eutectoid austenite at a testing temperature of 280° C. has a D.P.H. of 160+, which becomes 640 (with $P_u = 1000$ and $n = 2{\cdot}3$) at 100% bainite as quenched. Similarly, austenite at 245° C. has a diamond hardness of 200+, which becomes 700 (with $P_u = 1250$ and $n = 2{\cdot}37$) as transformed to 100% bainite and quenched.

In regard to the strain-hardening capacities of various steel microstructures, the stress–strain tensile diagram is now frequently expressed as $P = kl^m$, where m is the strain-hardening exponent. It has been suggested[10] that $n = 2 + m$, so that previous values for m published by one of the authors[8] would give $n = 2{\cdot}22$ for pearlite and 2·35 for austenite. Recent Japanese work,[15] however, has given a tensile value of $m = 0{\cdot}5$ for austenite, which agrees well with the present ball-indentation results and may be compared with n values for hard products obtained in the present work. Gensamer[16] also determined the tensile exponents for various structures, but bainites gave about the same values as pearlites. Elmendorf[17] found that certain combinations of martensite and bainite gave higher tensile strength and greater elongation values than either 100% bainite or 100% martensite, which is consistent with the present work.

Acknowledgments

Acknowledgment is made to the Council of The Iron and Steel Institute for a Carnegie Grant for one of the authors (G.R.B.), and to Mr. A. H. Moffatt, laboratory steward, who constructed the special apparatus used.

APPENDIX

The Hardness of Alloys of the System Fe–Fe_3C

By Hugh O'Neill

It has long been known that the indentation hardness of pearlitic plain-carbon steels bears a roughly linear relationship to their carbon content up to the eutectoid composition. Furthermore, the diamond hardness of the quenched steels up to 1·75% of carbon has been reported by various workers.[18–20] The hardness graph for the whole metastable system does not, however, seem to have been established, and so the literature was searched and steps were taken to obtain some necessary experimental values so as to outline the diagram.

The British Cast Iron Research Association supplied diamond pyramid results obtained on rolls of chilled iron and hardened steel; they also provided two pure alloys practically at the composition of the iron-carbide eutectic. Vickers tests on these specimens provided useful points on the graph.

Various values are available for cementite, and its diamond hardness was given as 820 by Lips and Sack.[21] Curve *A* in Fig. 7 thus represents the hardness diagram for the normal metastable conglomerate system; it is surprisingly linear, apart from the peak at the eutectoid composition. This peak is a general metallographic phenomenon and is smoothed out when the hardness of annealed and spheroidized steels is plotted. The cast eutectic alloy did not produce a peak.

In the quenched martensitic condition, the diamond pyramid hardness of wrought steels is generally considered to rise as a smooth steep curve to a maximum of about 1000 kg./sq. mm. for about 1·3% of carbon. At about 0·7% of carbon comes the complication of retained austenite, and quenching a 1·3% carbon steel above the A_{cm} line[20] gives a Rockwell hardness of *C*63, which may be plotted as 790 V.D.H. The specimen of white-iron eutectic in the cast condition gave a wide scatter of results, and, after heating at 1000° C. *in vacuo* for 1 hr. and water-quenching, the Vickers number was 930 max. The microstructure was not homogeneous, and so specimens were soaked for 7 hr. at 1000° C. and water-quenched. Vickers tests at 2·5 kg. gave a maximum value of 920, but some results as low as 870 were obtained. Subsequent heating at 50° C. for 2 hr. did not produce any definite effect. One area of one quenched specimen showed faint twin markings and had a hardness of 610 V.D.H. Pure cementite itself would not be expected to show any change of hardness on quenching, and, if this is

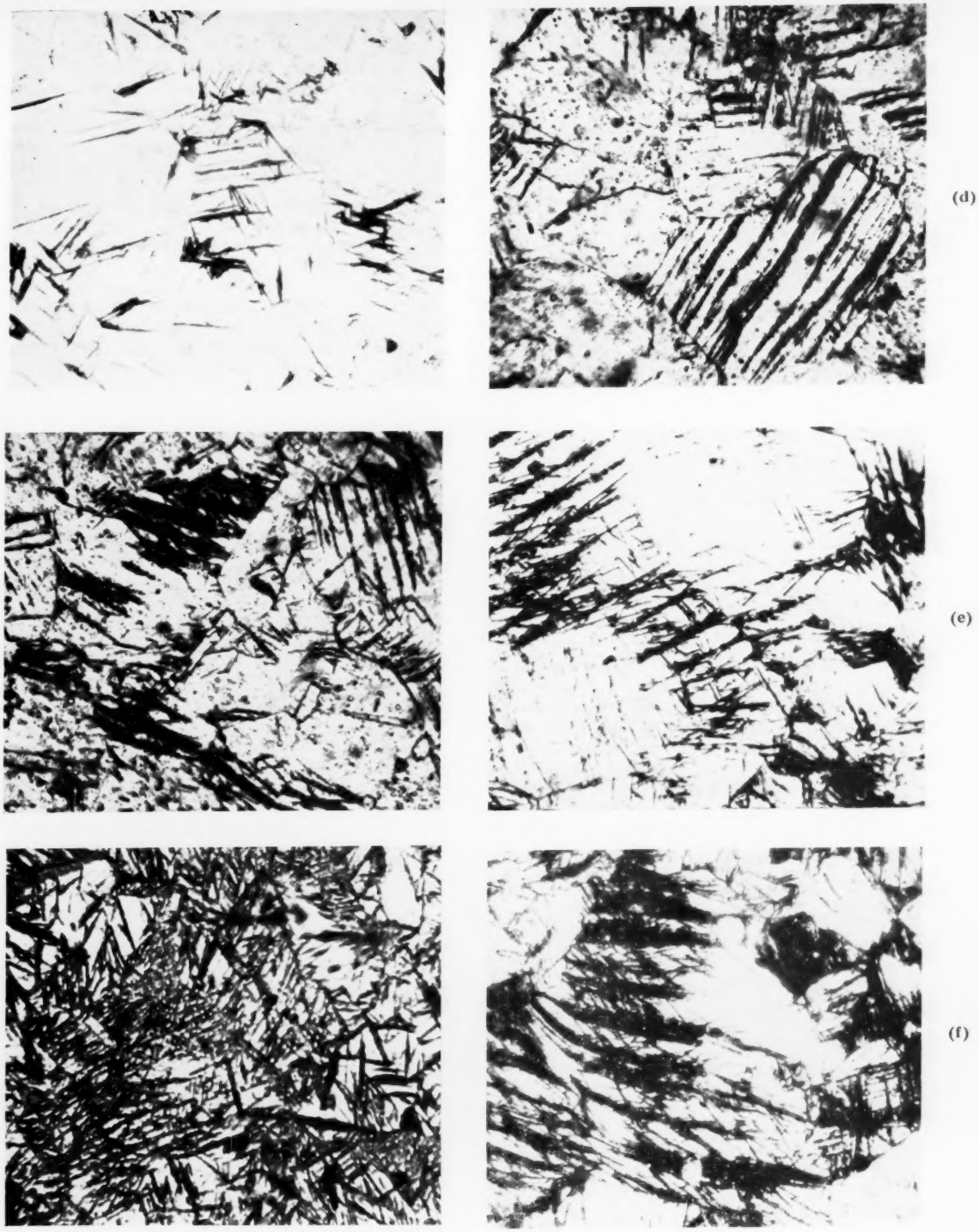


Fig. 6*a*—Held at 280° C. for 400 sec. Fig. 6*b* and *c*—Held at 280° C. for 490 sec. and deformed (*b*) 13%, (*c*) 41%

Fig. 6*d-f*—Held at 280° C. for 305 sec. and deformed (*d*) 13%, (*e*) 18%, (*f*) 58%

Fig. 6—Photomicrographs of quenched transformation products × 800

accepted, curve *B* is the graph for the homogeneous quenched alloys. In general, it rises through a maximum, and is therefore consistent with the diagram for a solid solution series.

The next step was to study the effect of cold-working both the normalized and the quenched alloys. Values of the former up to about 1% of carbon are available in the literature, and experiments on cementite had been made for the author during 1949 by Mr. P. Whitaker and Mr. L. H. Cope of Stewarts and Lloyds Ltd. Two samples of iron were carburized in different mixtures so as to produce massive cementite in the case. Hardness micro-tests were then made with the Stewarts and Lloyds instrument, after which the bars were rotated in a lathe and surface-deformed by traversing them slowly with a hemispherical diamond tool. Microscopic examination revealed some cracks in the distorted cementite, but pyramid micro-tests (50-g. load) showed that strain hardening had been effected. The results may be summarized as follows:

(*Iron basis material: Mn*, 0·40%; *Cu*, 0·012%; *Cr*, 0·016%)

	Carburized in: (i) Commercial mixture for 3 days at 1100° C.	(ii) Purified graphite; heated to melting in H.F. furnace and slowly cooled
Nitrogen (on surface layer), %	0·0125	0·0055
Average micro-hardness, kg./sq. mm.	1065	980
Micro-hardness after burnishing, kg./sq. mm.	>1367	1120
Increase, kg./sq. mm.	302	140

There is no reason to believe that the cementite has actually been strain-hardened to its maximum value, but it is safe to state that an increment of 300 may be obtained by cold-working. Results by Brown and Ineson[11] indicate that at this hardness level the Corby micro-tester gives values about 100–150 V.D.H. higher than the Vickers test at 2·5 kg., so that the average hardness of Fe_3C may be taken as 850 V.D.H., and 1150 V.D.H. at least may be reached by cold-working. Curve *C* represents the unquenched alloys after heavy cold-working.

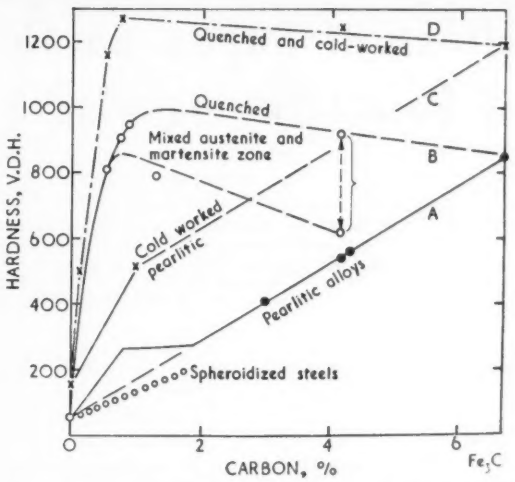


Fig. 7—Summarized hardness results on commercial iron–carbon alloys after various treatments

Superhardening with the Herbert cloudburst machine is one way of cold-working very hard metals. Herbert has published results[22] which indicate that by converting pendulum diamond hardness values to ball values, eutectoid normalized steel work-hardens to 520, and the quenched steel at 900 superhardens to 1275. The worked surfaces of pieces of quenched white-iron eutectic* gave Vickers values at 2·5-kg. load of 1246–1288 kg./sq. mm. These results enabled curve *D* to be drawn, in which the graph again passes through a maximum.

References

1. V. Biringuccio: "Pirotechnia" (translated by C. S. Smith and M. T. Gnudi), p. 371: 1942, New York, American Institute of Mining and Metallurgical Engineers.
2. D. Lewis: *J. Iron Steel Inst.*, 1929, No. I, pp. 427–441.
3. H. O'Neill: "The Hardness of Metals and Its Measurement": 1934, London, Chapman and Hall.
4. C. B. Post and W. S. Eberly: Reprint No. 5: 1946, Cleveland, American Society for Metals.
5. J. A. Jones: *The Metallurgist*, 1940, vol. 27, pp. 59–61.
6. N. P. Goss: *Trans. Amer. Soc. Metals*, 1945, vol. 34, pp. 630–638.
7. H. O'Neill, J. Jackson, and G. Farnham: *Phil. Mag.*, 1933, vol. 16, pp. 913–929.
8. H. O'Neill: *Proc. Inst. Mech. Eng.*, 1944, vol. 151, pp. 116–130.
9. W. Rostoker: *J. Inst. Metals*, 1950, vol. 77, pp. 175–184.
10. D. Tabor: "The Hardness of Metals": 1951, Oxford, Clarendon Press.
11. A. R. G. Brown and E. Ineson: *J. Iron Steel Inst.*, 1951, vol. 169, pp. 376–388.
12. M. Balicki: *Ibid.*, 1945, No. 1, pp. 181p–224p.
13. G. M. Enos, G. J. Peer, and J. C. Holtzwarth: *Metal Progress*, 1948, vol. 54, pp. 51–55.
14. M. D. Jepson and F. C. Thompson: *J. Iron Steel Inst.*, 1949, vol. 162, pp. 49–56.
15. S. Takeuchi and H. Suzuki: *Sci. Rep. Res. Inst. Tôhoku Univ.*, Ser. A, 1950, vol. 2, pp. 43–67.
16. M. Gensamer: *Trans. Amer. Soc. Metals*, 1946, vol. 36, pp. 30–60.
17. H. J. Elmendorf: *Ibid.*, 1944, vol. 33, pp. 236–255.
18. H. Esser and W. Eilender: *Arch. Eisenhüttenwesen*, 1930, vol. 4, pp. 113–144.
19. E. C. Bain: "The Functions of the Alloying Elements in Steel": 1939, Cleveland, American Society for Metals.
21. E. M. H. Lips and J. Sack: *Nature*, 1936, vol. 138, pp. 328–329.
20. K. L. Clark and N. Kowall: *Trans. Amer. Inst. Min. Met. Eng.*, 1944, vol. 158, pp. 328–332; *Metals Technology*, 1944, vol. 11, T.P. No. 1681.
22. E. G. Herbert: *J. Iron Steel Inst.*, 1927, No. II, pp. 265–282; *Trans. Amer. Soc. Steel Treating*, 1929, vol. 16, pp. 77–96; *see also* O'Neill.[3]

* Shot-peened by Tilghman's Patent Sand Blast Co. Ltd.

Effect of Boron on the Mechanical Properties of Low-Alloy Steels

By R. Wilcock, A.Met., A.I.M.

SYNOPSIS

Five low-alloy steels have been made, each with one ingot treated with boron and one ingot without boron. Tensile and impact tests in the hardened and tempered condition have been carried out in a range of sizes up to 4 in. sq., and the hardenability has been investigated by Jominy end-quench tests and hardness traverses. From these data, an attempt has been made to assess the possibilities and extent of alloy conservation that can be achieved by boron treatment on the basis of typical British specifications. 757

THE manufacture of boron-treated, low-alloy, heat-treatable steels in the U.S.A. has now reached some magnitude, and there is considerable practical experience there of the use of these steels in service. It is claimed that appreciable savings in strategic metals, such as Ni and Mo, have been made by replacing relatively highly alloyed steels by boron-treated steels of lower alloy content.

In the application of these steels, however, almost the only criterion of suitability has been the hardenability as assessed by the Jominy test, and usually the substitution has taken place on the basis of roughly equal hardenability in the section being considered. For this reason the published data available on these steels, both in the U.S.A. and in Britain, give very little information of the type required by British designers on the mechanical properties available in various ruling sections; nor do they enable a comparison to be made between boron-treated and boron-free steels of the same compositions.

To provide this type of information, and to compare it with hardenability data on a few steels, five experimental, 12-cwt., high-frequency furnace heats have been made, each poured into two ingots, the first not treated and the second treated with boron. The compositions were as follows:

Type of Steel	C, %	Mn, %	Ni, %	Cr, %	Mo, %
En 15	0·35	1·45	...	...	...
	0·35	1·45	...	...	0·10
	0·35	1·45	...	...	0·20
En 16	0·35	1·45	...	...	0·28
NE 8640 (modified)	0·40	0·90	0·55	0·40	0·12

The first four casts make up a series based on a C–Mn steel of the En 15 type; succeeding heats had Mo added in increasing amounts, the limiting composition being of the En 16 type. From this series it was hoped to assess the effect of boron in various sizes, and to show whether boron could replace Mo, an alloying element in short supply. One heat was made of the modified American NE 8640 type with reduced Mo, this being included as typical of the multiple-low-alloy boron steels so extensively used in the U.S.A. as substitutes for higher alloyed compositions.

EXPERIMENTAL PROCEDURE

Manufacture of the Steel

To avoid, as far as possible, any variation in residual elements, sufficient raw material was put aside to manufacture all the steels. The mixtures were primarily of basic open-hearth billets with additions of pure pig iron to adjust the carbon contents. The analyses of these raw materials were as follows:

	O.H. Billets	Pig Iron
C, %	0·08	3·60
Mn, %	0·22	0·3
Si, %	0·17	0·2
S, %	0·027	0·025
P, %	0·013	0·030
Ni, %	0·11	tr.
Cr, %	0·03	tr.
Mo, %	0·03	tr.

The steels were manufactured in a basic-lined, 12-cwt., high-frequency induction furnace and were cast into 8-in. sq., 6-cwt. ingots. After melting the billets and pig iron, the necessary amounts of ferro-silicon and ferro-manganese were added, followed, if required, by Ni, Cr, or Mo; then 1 lb. of calcium silicide was added to the slag. Immediately before tapping, and at a temperature of 1630–1640° C., an addition of 16 oz. of Al per ton of steel was made, followed by the tundish pouring of the first of the two ingots. The slag-free metal remaining in the furnace was then treated with 4·2 lb. of a complex boron alloy per ton of steel, and was cast almost immediately, thus giving two ingots which, apart from the boron treatment, were as nearly identical as possible.

The complex boron alloy had a composition of 1·85% of boron, 25% of Al, 29% of Ti, and 9·5% of Si. The additions were intended to yield approximately 0·0040% of boron, with 0·06% of Al and 0·065% of Ti. The yield of boron was in all cases good, ranging from 0·0030 to 0·0041%, and being present substantially in the desirable acid-soluble form recommended by Bardgett and Reeve.[1] There were indications of slight boron contamination of the non-boron parts of the melts, probably arising from erosion of the furnace lining, as it occurred in those heats which did not follow boron-treated heats. The figures for boron quoted in the Tables were obtained on test ingots taken at the time of casting, but these were in close agreement with estimations on the bar material. With the exception of carbon content, the analyses of the two ingots from any one cast are very similar. The inevitable slight delay between casting the two ingots tended to give a slight drop in carbon; several heats had to be discarded,

Manuscript received 16th February, 1953.

Mr. Wilcock is Works Metallurgist at Messrs. Samuel Fox and Co., Ltd., Stocksbridge.

Table I

CARBON–MANGANESE STEEL (En 15)

	C, %	Mn, %	Si, %	S, %	P, %	Ni, %	Cr, %	Mo, %	B, % Soluble	B, % Insol.	B, % Total	Al, %	Ti, %	Grain Size
5380 (NB)	0·37	1·53	0·24	0·027	0·020	0·13	0·03	0·03	0·0002	0·0002	0·0004	0·06	0·01	7–8
5380 (B)	0·36	1·49	0·27	0·027	0·018	0·13	0·03	0·03	0·0027	0·0003	0·0030	0·10	0·06	8

Hardened in oil from 850° C.

Size, in., and Position*	Tempering Temp., C.	Boron (B) or No Boron (NB)	0·1% Proof Stress	0·2% Proof Stress	0·5% Proof Stress	Yield Stress	Ultimate Stress	Elongation, %	Reduction in Area, %	Izod Impact, ft.lb.
			Tons/sq. in.							
0·424 dia.	250	NB	96·5	100·9	106·7		121·4	14·7	43·3	12
	250	B	93·1	97·8	105·7		117·2	14·7	46·8	22
	300	NB	96·2	97·9	100·7		110·4	14·7	46·8	9
	300	B	97·2	98·9	102·5		111·8	14·7	49·6	17
	350	NB	76·9	82·9	87·2		92·3	13·3	39·7	6
	350	B	91·8	93·3	95·6		103·1	16·0	49·6	14
	400	NB	83·6	84·4	85·9		91·6	16·0	56·7	9
	400	B	84·3	84·8	85·9		89·9	14·7	54·6	14
	450	NB	71·2	71·4	71·7		77·2	17·3	53·2	44
	450	B	74·5	74·6	74·7		76·2	16·7	56·7	46
	500	NB	62·9	63·1	63·3		68·4	25·3	65·2	58
	500	B	62·6	63·1	63·6		69·6	21·3	56·7	57
	550	NB	58·5	58·7	58·9		64·5	22·7	63·8	63
	550	B	57·8	57·9	58·2		63·9	22·7	59·6	66
	600	NB	48·9	49·0	49·1		54·6	24·7	66·7	79
	600	B	50·1	50·2	50·4		54·6	24·7	61·0	77
0·564	200	NB				78·8	85·2	8·5	24·4	10
dia.	200	B				107·6	122·8	12·0	36·4	23
¾ dia.	500	NB				43·8	55·2	23·0	59·2	69
	500	B				65·1	68·2	19·0	57·2	51
	550	NB				41·9	52·7	25·5	61·6	81
	550	B				58·5	61·9	22·0	57·2	58
1⅛ dia.	500	NB				42·7	53·6	24·0	61·6	76
	500	B				63·2	68·5	18·0	57·2	56
	550	NB				38·9	52·3	25·5	57·2	77
	550	B				58·8	63·3	19·0	59·2	59
	600	NB				37·1	48·6	29·0	66·0	87
	600	B				51·8	56·8	23·0	61·6	75
1½ dia.	550	B				48·0	55·7	21·0	59·2	63
1¾ dia.	550	B				43·8	53·8	23·0	59·2	51
2 dia.	450	B				48·7	60·4	18·5	49·6	13
	500	NB				39·8	53·0	27·0	61·6	78
	500	B				44·7	55·4	22·0	59·2	31
	550	NB				38·8	51·4	27·0	63·6	81
	550	B				39·2	53·2	24·0	59·2	34
	600	NB				37·5	48·3	30·0	66·0	87
	600	B				35·5	48·7	26·0	61·6	49
2½ dia.	500	NB				41·9	53·1	26·0	63·6	72
	500	B				39·3	53·8	22·0	57·2	33
	550	NB				40·4	51·4	29·0	63·6	72
	550	B				35·2	50·3	24·0	59·2	41
	600	NB				35·3	48·7	29·0	63·6	79
	600	B				34·3	49·8	25·0	59·2	57
4 sq. M	500	NB				40·8	51·8	26·5	59·2	68
M	500	B				34·4	50·2	23·0	54·8	18
O	500	NB				39·3	51·8	27·5	59·2	69
O	500	B				36·4	52·4	23·0	57·2	23
M	550	NB				36·4	51·0	27·0	59·2	71
M	550	B				33·2	48·6	26·0	54·8	24
O	550	NB				36·9	50·4	27·0	57·2	71
O	550	B				35·4	50·2	23·0	54·8	27
M	600	NB				34·4	47·6	29·0	61·6	73
M	600	B				35·3	49·1	24·0	52·4	22
O	600	NB				34·6	48·3	29·0	63·6	73
O	600	B				33·9	49·9	26·0	59·2	31
	650	B				29·9	44·8	27·0	57·2	48

* M = midway between axis and outside surface, O = outside surface.

but in those retained the difference is marked only in cast *B*5382.

Forging and Treatment of the Steel

The ingots were hammer-forged to 4 in. sq., and the whole range of test pieces required was taken from this size by sawing and subsequent forging to provide the smaller sections.

The steels were all quenched in oil from 850° C. in the routine laboratory manner. With the exception of the treatments at 200° C., which were followed by air-cooling, oil-quenching was used after tempering. Soaking times for hardening varied from 1 to 2½ hr., depending on the section, but all specimens were tempered for 3 hr.

Most of the work was done using sizes of 1⅛-, 2-, and 2½-in. dia. and 4 in. sq., all except the 2-in. dia. being used as a basis for the mechanical tests specified for the En series of steels in B.S. 970. Tempering temperatures of 500°, 550°, and 600° C. were used, it being considered that 500° C. represented the minimum temperature likely to be adopted in normal commercial practice. Tests in the very lightly tempered condition were heat-treated in test-piece sizes only, and on two casts a more extended series of tempering temperatures was used, the treatment again being carried out in test-piece sizes.

Mechanical Properties

Tensile and notched-bar impact tests were carried out on each steel treated in various sizes. Up to and including 2 in. dia., the test pieces were machined from the centres of the bars; the 2½-in. dia. pieces were taken from a position midway between the axis and outside surface, and the 4-in. dia. pieces from both outside and midway positions.

In most cases, yield-stress values were determined with dividers, but proof-stress values were recorded for certain tests on two of the steels.

A limited number of impact tests were carried out to determine the effect of temperature of testing, the specimens being treated in 1⅛ in. dia. and tempered

Table II

MANGANESE–MOLYBDENUM STEEL (En 15)

	C, %	Mn, %	Si, %	S, %	P, %	Ni, %	Cr, %	Mo, %	B, % Soluble	B, % Insol.	B, % Total	Al, %	Ti, %	Grain Size
5513 (NB)	0·33	1·40	0·20	0·026	0·019	0·11	0·03	0·11	0·0001	0·0002	0·0003	0·055	0·01	7–8
5513 (B)	0·32	1·39	0·24	0·029	0·018	0·12	0·05	0·11	0·0040	0·0001	0·0041	0·04	0·05	7–8

Hardened in oil from 850° C.

Size, in., and Position*	Tempering Temp., ° C.	Boron (B) or No Boron (NB)	0·1% Proof Stress	0·2% Proof Stress	0·5% Proof Stress	Yield Stress	Ultimate Stress	Elongation, %	Reduction in Area, %	Izod Impact, ft.lb.
			Tons/sq. in.							
0·424 dia.	200	NB	81·6	89·0	99·7		120·35	7·3	22·4	19
	200	B	82·9	89·4	98·0		115·25	14·0	43·3	19
1⅛ dia.	500	NB				43·3	53·4	23·5	61·6	77
	500	B				63·7	66·7	19·0	59·2	45
	550	NB				41·9	52·9	24·5	61·6	79
	550	B				62·0	64·9	21·0	61·6	56
	600	NB				39·0	48·9	27·0	63·6	85
	600	B				51·8	56·4	22·0	63·6	71
2 dia.	500	NB				42·6	52·0	25·0	61·6	84
	500	B				50·1	57·6	22·0	59·2	31
	550	NB				39·0	49·4	26·0	63·6	85
	550	B				46·2	53·0	22·0	59·2	38
	600	NB				36·4	46·0	28·0	66·0	94
	600	B				41·6	49·8	24·0	63·6	69
2½ dia.	500	NB				41·0	51·8	25·0	63·6	74
	500	B				43·8	52·9	21·5	61·6	28
	550	NB				37·2	49·9	26·0	61·6	79
	550	B				41·3	51·9	23·0	59·2	37
	600	NB				35·9	47·2	28·0	66·0	87
	600	B				38·1	48·9	26·0	63·6	44
4 sq. M	500	NB				40·9	50·0	23·5	59·2	59
M	500	B				40·8	51·1	24·0	57·2	27
O	500	NB				41·1	49·9	24·5	61·6	66
O	500	B				41·8	50·7	24·0	61·6	22
M	550	NB				36·1	48·0	25·5	57·2	63
M	550	B				37·6	49·2	25·0	59·2	29
O	550	NB				37·4	48·3	24·0	59·2	64
O	550	B				37·8	49·9	23·0	59·2	26
M	600	NB				34·7	47·0	25·0	59·2	77
M	600	B				36·7	46·8	25·0	61·6	34
O	600	NB				34·4	46·8	25·0	61·6	77
O	600	B				38·8	47·3	26·0	61·6	32

* M = midway between axis and outside surface, O = outside surface.

at 550° C.; except for these, all other mechanical tests were carried out at 19° C.

The McQuaid-Ehn grain size was determined for each steel in accordance with the A.S.T.M. method.

End-Quenched Hardenability Curves

The hardenability curves were determined on specimens machined in all cases from 1¼-in. dia. bar end-quenched from 850° C., the procedure being in accordance with that laid down by the T.A.C. Hardenability Sub-Committee.[2]

In addition to those tested in the as-quenched condition, other specimens were tested after tempering at 500°, 550°, and 600° C. However, since the curves were all very similar in form, the results at 550° C. have not been reproduced.

Hardness Traverse Data

Surveys of the hardness across samples 1-, 2-, 3-, and 4-in. dia., oil-quenched in the laboratory, were carried out using the Rockwell C and Rockwell B hardness testing methods. As the curves were intended to give only a graphical impression of the hardness achieved, the figures have been converted to the more commonly used Brinell hardness numbers. In an earlier work,[3] hardness results in the as-quenched condition on such traverses were rather erratic, and this was also the case in a few of the sections tested here, particularly those at the borderline between good and poor hardening. In spite of this, no actual points have been recorded on the curves, which, as emphasized earlier, are included not to give any absolute values, but merely to indicate the effectiveness of quenching under laboratory conditions.

DISCUSSION OF RESULTS

The results of mechanical tests carried out at 19° C. are quoted in Tables I–V. The Jominy end-quenched hardenability curves and hardness traverses are reproduced in Figs. 1–5. Notched-bar impact tests at reduced temperatures are recorded in Table VI.

In attempting to assess the influence of boron on the mechanical properties it is convenient to consider

Table III

MANGANESE–MOLYBDENUM STEEL (En 15)

	C, %	Mn, %	Si, %	S, %	P, %	Ni, %	Cr, %	Mo, %	B, % Soluble	B, % Insol.	B, % Total	Al, %	Ti, %	Grain Size
5609 (NB)	0·38	1·44	0·30	0·028	0·017	0·13	0·07	0·21	0·0002	0·0001	0·0003	0·02	0·01	7-8
5609 (B)	0·36	1·43	0·33	0·030	0·017	0·14	0·06	0·20	0·0038	0·0001	0·0039	0·06	0·045	7-8

Hardened in oil from 850° C.

Size, in., and Position*	Tempering Temp., °C.	Boron (B) or No Boron (NB)	0·1% Proof Stress	0·2% Proof Stress	0·5% Proof Stress	Yield Stress	Ultimate Stress	Elongation, %	Reduction in Area, %	Izod Impact, ft.lb.
			Tons/sq. in.							
0·424 dia.	200	NB	85·1	92·2	102·3		125·6	9·3	27·7	10
	200	B	87·2	93·9	103·9		124·1	12·0	41·1	17
1⅛ dia.	500	NB				56·2	62·9	19·0	52·4	46
	500	B				74·0	76·9	18·0	57·2	43
	550	NB				49·2	59·2	21·0	52·4	52
	550	B				69·8	73·8	19·0	54·8	49
	600	NB				47·3	55·9	23·0	61·6	75
	600	B				58·7	63·7	23·0	59·2	66
2 dia.	500	NB				48·4	58·4	21·0	52·4	43
	500	B				65·7	69·9	16·0	47·2	23
	550	NB				46·0	56·0	21·5	57·2	58
	550	B				60·7	67·2	17·0	52·4	33
	600	NB				42·3	52·0	25·0	61·6	71
	600	B				53·7	60·8	20·0	57·2	55
2½ dia.	500	NB				50·6	57·9	21·0	52·4	41
	500	B				61·6	67·2	15·0	47·2	24
	550	NB				46·8	57·6	22·0	57·2	61
	550	B				55·4	61·8	19·0	52·4	32
	600	NB				42·4	52·8	25·0	59·2	71
	600	B				49·9	57·4	21·5	57·2	48
4 sq. M	500	NB				48·6	57·8	17·0	36·4	20
M	500	B				54·3	60·5	14·0	33·6	11
O	500	NB				46·3	57·1	17·0	39·2	28
O	500	B				52·6	59·8	15·0	33·6	11
M	550	NB				47·2	56·0	19·0	44·4	29
M	550	B				50·0	57·8	19·0	47·2	16
O	550	NB				45·2	55·0	22·0	49·6	45
O	550	B				51·6	58·0	19·0	49·6	14
M	600	NB				40·0	52·0	22·0	52·4	50
M	600	B				43·9	54·8	21·0	52·4	21
O	600	NB				42·3	51·4	22·0	52·4	54
O	600	B				46·1	55·3	21·0	54·8	18

* M = midway between axis and outside surface, O = outside surface.

the results in size groups, as mass effect has a profound influence on the inter-relationship of tensile and impact properties. These size groups—the limits of which vary from steel to steel—are as follows:

Small Sizes—Where both boron-free and boron-treated parts are substantially hardened through
Intermediate Sizes—Where the boron-free steel does not harden fully but the boron-treated steel does
Large Sizes—Where neither steel hardens fully.

The mechanical test results fall fairly clearly into these various groups, which can also be related to the end-quench hardenability curves discussed on p. 417.

Small Sizes

The results quoted in the Tables for test pieces heat-treated in 0·424-in. dia. and tempered at 200° or 250° C. show generally similar ultimate stresses, with a tendency for the boron-treated parts to be slightly lower than the untreated. The effect on the impact values is pronounced, showing a marked improvement with the addition of boron. The values for ductility, as expressed by elongation and reduction in area, are rather erratic, but the pattern is fairly clear, the addition of boron being generally beneficial. On two steels complete tempering curves have been produced in 0·424-in. dia., and one of these is reproduced in Fig. 6. Ultimate-stress values are very similar for both boron-free and boron-treated parts of the same cast for a given tempering treatment. The boron-treated parts maintain their advantage in impact properties over the non-boron parts at all treatments up to 400° C.; at 450° C. and higher tempering temperatures, however, the impacts of the two halves of each steel are similar, showing that the advantage held by the boron treatment at the lower tempering temperatures has been lost.

The results on the C–Mn steel, cast 5380, are interesting. For the intensity of quench used, the limit for full hardening of the boron-free steel is about $\frac{1}{2}$ in. dia. Thus, while both parts of this cast hardened fully in 0·424 in. dia., only the boron-treated part did so in 0·564 in. dia. In the latter size the boron steel had

Table IV

MANGANESE–MOLYBDENUM STEEL (En 16)

	C, %	Mn, %	Si, %	S, %	P, %	Ni, %	Cr, %	Mo, %	B, % Soluble	B, % Insol.	B, % Total	Al, %	Ti, %	Grain Size
5514 (NB)	0·35	1·44	0·26	0·028	0·020	0·14	0·07	0·28	0·0001	0·0001	0·0002	0·065	0·01	7–8
5514 (B)	0·33	1·44	0·29	0·030	0·019	0·15	0·06	0·28	0·0034	0·0002	0·0036	0·08	0·055	7–8

Hardened in oil from 850° C.

Size, in., and Position*	Tempering Temp., °C.	Boron (B) or No Boron (NB)	0·1% Proof Stress	0·2% Proof Stress	0·5% Proof Stress	Yield Stress	Ultimate Stress	Elongation, %	Reduction in Area, %	Izod Impact, ft.lb.
			Tons/sq. in.							
0·424 dia.	200	NB	86·8	93·5	102·9		122·9	13·3	39·7	16
	200	B	82·9	91·6	101·2		118·4	14·0	39·7	22
1⅛ dia.	500	NB				70·8	72·6	17·5	52·4	53
	500	B				78·0	79·9	16·0	49·6	35
	550	NB				59·6	65·2	19·0	57·2	59
	550	B				71·6	72·8	17·0	54·8	53
	600	NB				51·0	57·9	23·0	61·6	72
	600	B				62·4	64·6	17·5	57·2	56
2 dia.	500	NB				56·2	61·1	19·0	57·2	61
	500	B				69·5	73·0	16·0	52·4	18
	550	NB				51·2	58·1	22·0	57·2	69
	550	B				63·7	68·9	18·0	52·4	54
	600	NB				42·4	52·4	24·0	61·6	82
	600	B				55·2	60·8	21·0	59·2	65
2½ dia.	500	NB				52·7	59·8	19·0	54·8	63
	500	B				65·8	70·6	18·0	49·6	21
	550	NB				49·2	57·2	22·0	57·2	69
	550	B				61·3	65·6	18·0	54·8	37
	600	NB				43·4	52·8	24·0	63·6	79
	600	B				50·8	58·1	21·0	57·2	63
4 sq. M	500	NB				55·7	61·8	13·0	24·4	14
M	500	B				56·3	61·2	16·0	39·2	24
O	500	NB				54·3	62·2	17·0	42·0	24
O	500	B				57·4	62·8	18·0	49·6	14
M	550	NB				52·0	58·6	17·0	44·4	44
M	550	B				51·8	59·1	20·0	52·4	28
O	550	NB				51·2	58·9	19·5	54·8	35
O	550	B				53·4	59·5	18·5	52·4	24
M	600	NB				43·3	52·8	21·5	52·4	71
M	600	B				43·8	52·7	22·0	54·8	44
O	600	NB				46·0	54·9	20·0	54·8	66
O	600	B				44·8	53·4	22·0	59·2	39

* M = midway between axis and outside surface, O = outside surface.

Table V

NICKEL–CHROMIUM–MOLYBDENUM STEEL (NE 8640 MODIFIED)

	C, %	Mn, %	Si, %	S, %	P, %	Ni, %	Cr, %	Mo, %	B, % Soluble	B, % Insol.	B, % Total	Al, %	Ti, %	Grain Size
5382 (NB)	0·40	0·91	0·31	0·024	0·018	0·55	0·43	0·14	0·0002	0·0001	0·0003	0·06	0·01	7–8
5382 (B)	0·37	0·91	0·33	0·025	0·016	0·55	0·42	0·14	0·0031	0·0002	0·0033	0·07	0·035	8

Hardened in oil from 850° C.

Size, in., and Position*	Tempering Temp., °C.	Boron (B) or No Boron (NB)	0·1% Proof Stress	0·2% Proof Stress	0·5% Proof Stress	Yield Stress	Ultimate Stress	Elongation, %	Reduction in Area, %	Izod Impact, ft.lb.
			Tons/sq. in.							
0·424 dia.	250	NB	98·3	104·1	112·1		125·2	11·3	31·9	14
	250	B	95·7	103·8	110·6		120·6	16·0	46·8	18
	300	NB	101·8	102·7	106·4		115·8	13·3	44·7	7
	300	B	102·8	104·1	106·4		116·9	12·7	43·3	16
	350	NB	97·0	97·9	101·9		109·9	14·0	43·3	11
	350	B	98·6	99·0	100·7		107·0	12·7	46·8	15
	400	NB	93·3	93·8	95·0		100·2	15·3	44·7	15
	400	B	90·4	91·2	91·6		97·3	15·3	46·8	15
	450	NB	82·1	82·6	83·7		87·9	15·3	53·2	37
	450	B	81·9	82·3	82·6		87·3	16·7	53·2	31
	500	NB	73·8	74·5	75·1		79·1	16·7	59·6	42
	500	B	72·2	72·3	72·6		77·6	17·3	49·6	42
	550	NB	62·9	63·1	63·6		71·8	19·3	59·6	53
	550	B	70·1	70·2	70·4		72·7	23·3	59·6	54
	600	NB	57·4	57·5	57·6		64·5	24·7	62·4	66
	600	B	59·5	59·6	59·8		63·5	22·0	59·6	66
0·564 dia.	200	NB				108·4	129·6	6·0	8·4	16
	200	B				119·6	127·2	12·5	36·4	29
1⅛ dia.	500	NB				67·4	70·8	17·5	57·2	55
	500	B				77·4	79·4	16·0	47·2	41
	550	NB				59·7	67·0	23·0	59·2	64
	550	B				69·8	73·0	17·0	54·8	53
	600	NB				49·4	58·6	22·0	61·6	82
	600	B				61·6	65·2	20·0	57·2	62
1½ dia.	550	NB				49·6	59·1	21·0	57·2	67
	550	B				64·0	69·8	19·0	54·8	43
1¾ dia.	550	NB				46·4	58·0	22·0	57·2	71
	550	B				61·9	65·0	19·5	54·8	46
2 dia.	450	B				77·8	81·1	14·0	42·0	15
	500	NB				57·6	62·8	19·5	57·2	63
	500	B				71·0	75·8	17·0	49·6	29
	550	NB				46·8	58·0	23·0	59·2	67
	550	B				66·3	71·2	17·5	47·2	34
	600	NB				42·8	53·8	24·5	61·6	79
	600	B				57·2	63·0	20·0	59·2	59
2½ dia.	500	NB				53·9	60·8	19·0	52·4	64
	500	B				62·9	68·8	15·0	42·0	21
	550	NB				46·9	57·8	21·0	54·8	67
	550	B				59·6	65·2	18·0	52·4	38
	600	NB				41·2	52·7	24·0	63·6	76
	600	B				50·2	59·4	11·0	57·2	59
4 sq. M	500	NB				50·7	59·0	19·0	42·0	37
M	500	B				55·6	62·8	17·0	42·0	18
O	500	NB				52·0	58·0	21·0	54·8	54
O	500	B				60·2	65·6	17·5	52·4	18
M	550	NB				42·8	55·2	19·5	47·2	53
M	550	B				52·9	57·2	16·0	39·2	31
O	550	NB				43·9	55·4	20·0	52·4	60
O	550	B				53·2	59·6	17·0	47·2	25
M	600	NB				41·5	51·4	23·0	57·2	65
M	600	B				46·9	56·1	20·5	49·6	36
O	600	NB				40·2	51·6	22·0	54·8	70
O	600	B				49·2	58·4	20·0	54·8	35

* M = midway between axis and outside surface, O = outside surface.

Table VI

EFFECT OF TESTING TEMPERATURE ON NOTCHED-BAR IMPACT VALUES

Bars 1¼-in. dia., oil-quenched 850° C., tempered 550° C.

Cast	Type of Steel	Boron (B) or No Boron (NB)	Izod Impact, ft.lb., Tested at:			
			19° C.	0° C.	− 20° C.	− 40° C.
5380	1¼% Mn, 0·03% Mo	NB	74, 80, 77	93, 82	75, 84	80, 75
		B	58, 59, 61	63, 65	54, 36	33, 22
5513	1¼% Mn, 0·11% Mo	NB	80, 84, 73	86, 96	92, 84	82, 80
		B	58, 56, 54	52, 41	38, 35	29, 23
5609	1¼% Mn, 0·20% Mo	NB	52, 49, 51	37, 50	21, 19	18, 18
		B	48, 49, 51	31, 32	29, 36	19, 19
5514	1¼% Mn, 0·28% Mo	NB	61, 61, 55	62, 65	64, 62	48, 45
		B	54, 49, 55	47, 50	42, 46	31, 32
5382	NE 8640	NB	66, 65, 62	66, 65	59, 60	61, 64
		B	54, 53, 52	47, 50	37, 37	30, 26

thus a marked advantage in ultimate stress, ductility, and impact values.

Intermediate Sizes

Except for a few tests on cast 5380 in ¾-in. dia., the next size larger than the test-piece sizes is 1⅛-in. dia., and under the quenching condition used, even the most highly alloyed steels investigated did not harden fully in this size when boron was not added, although two of them were evidently just short of full hardening. This is illustrated by the following figures, which show that the increase in ultimate stress caused by the addition of boron is less for the two more highly alloyed steels:

	Steel	Mo, %	Increase in U.T.S., tons/sq. in.
Bars 1½-in. dia., O.Q. 850° C., T. 550° C.	Mn–Mo	0·03	11
	,,	0·11	12
	,,	0·20	15
	,,	0·28	7
	NE 8640	0·14	6

With this increase in tensile strength there is a drop in impact and ductility. It is therefore not possible to compare the two parts of the same steels that received the same tempering treatments, but if the boron steels are tempered at higher temperatures to bring the ultimate-stress values down to the level of the untreated steels, the impact and ductility values are raised accordingly, so that there is then, for all

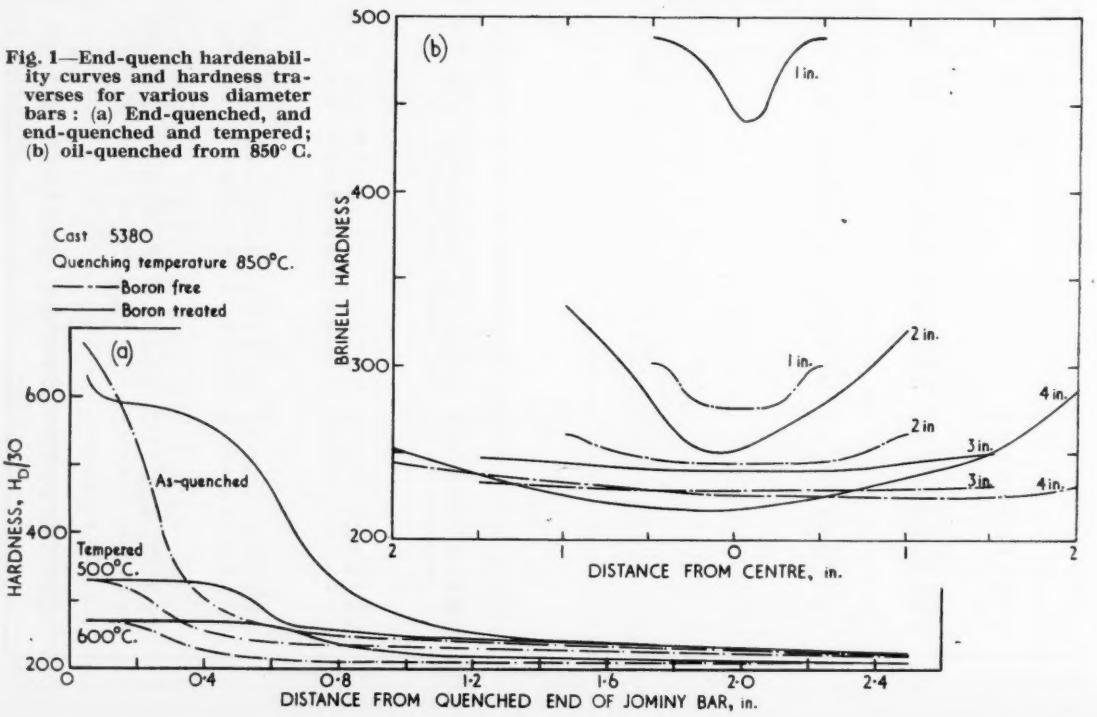


Fig. 1—End-quench hardenability curves and hardness traverses for various diameter bars: (a) End-quenched, and end-quenched and tempered; (b) oil-quenched from 850° C.

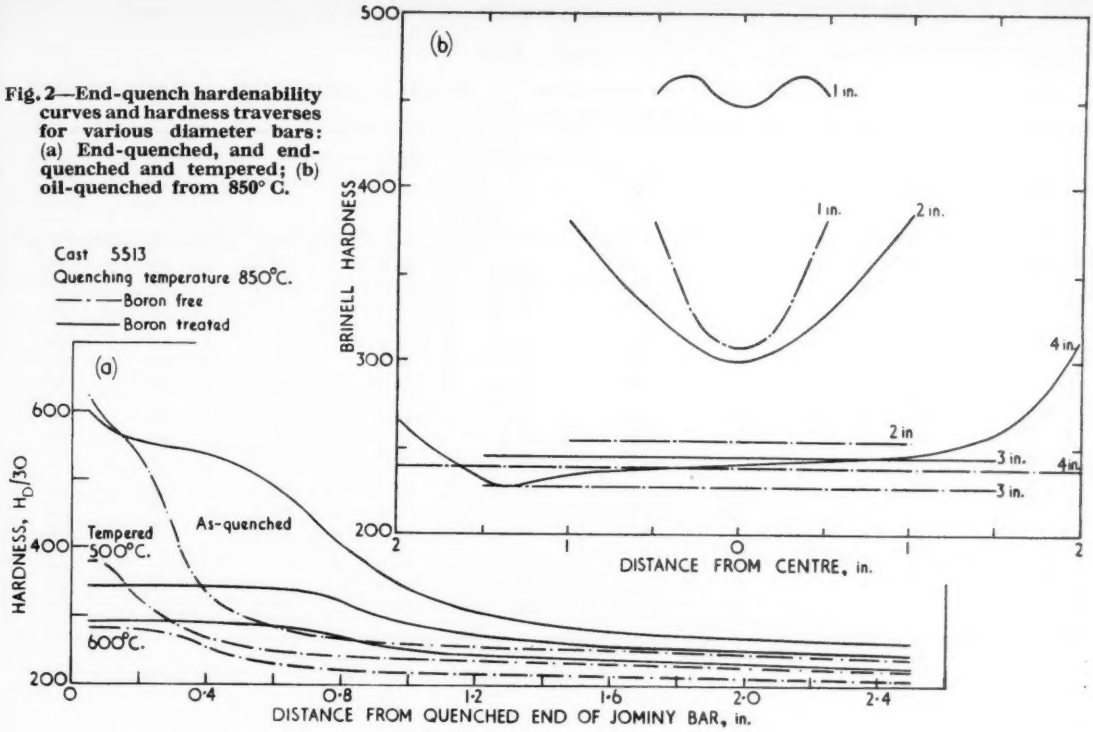


Fig. 2—End-quench hardenability curves and hardness traverses for various diameter bars: (a) End-quenched, and end-quenched and tempered; (b) oil-quenched from 850° C.

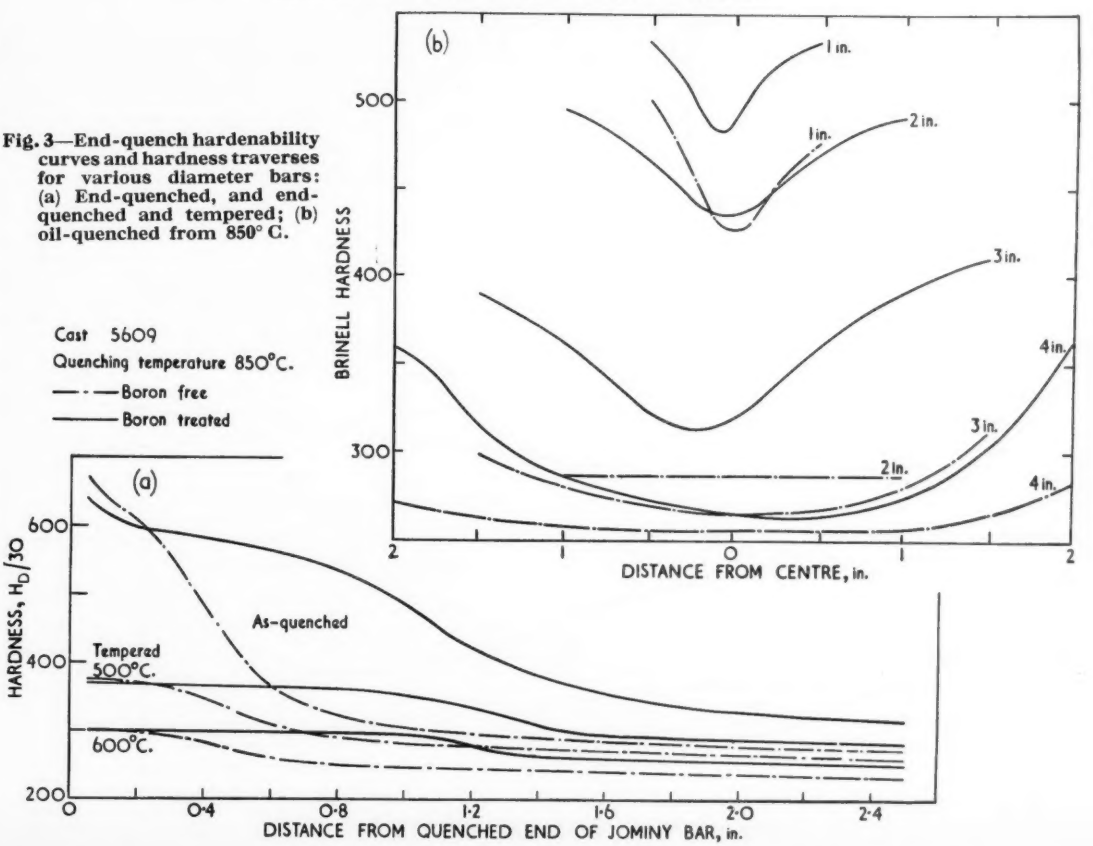


Fig. 3—End-quench hardenability curves and hardness traverses for various diameter bars: (a) End-quenched, and end-quenched and tempered; (b) oil-quenched from 850° C.

Table VII

EQUALIZATION OF MECHANICAL PROPERTIES BY VARYING TEMPERING TEMPERATURES

Bars 1¼-in. dia., oil-quenched 850° C.

Cast	Type of Steel	Boron (B) or No Boron (NB)	Tempering Temp., °C.	Ultimate Stress (Tons/sq. in.)	Yield Stress (Tons/sq. in.)	Elongation, %	Reduction in Area, %	Izod Impact ft.lb.
5380	1½% Mn, 0·03% Mo	NB	550	52·3	38·9	25·5	57	77
		B	550	63·3	58·8	19·0	59	59
		B	600	56·8	51·8	23·0	62	75
5513	1½% Mn, 0·11% Mo	NB	550	52·9	41·9	24·5	62	79
		B	550	64·9	62·0	21·0	62	56
		B	600	56·4	51·8	22·0	64	71
5609	1½% Mn, 0·20% Mo	NB	550	59·2	49·9	21·0	52	52
		B	550	73·8	69·8	19·0	55	49
		B	600	63·7	58·7	23·0	59	66
5514	1½% Mn, 0·28% Mo	NB	550	65·2	59·6	19·0	57	59
		B	550	72·8	71·6	17·0	55	53
		B	600	64·6	62·4	17·5	57	56
5382	NE 8640	NB	550	67·0	59·7	23·0	59	64
		B	550	73·0	69·8	17·0	55	53
		B	600	65·2	62·6	20·0	57	62

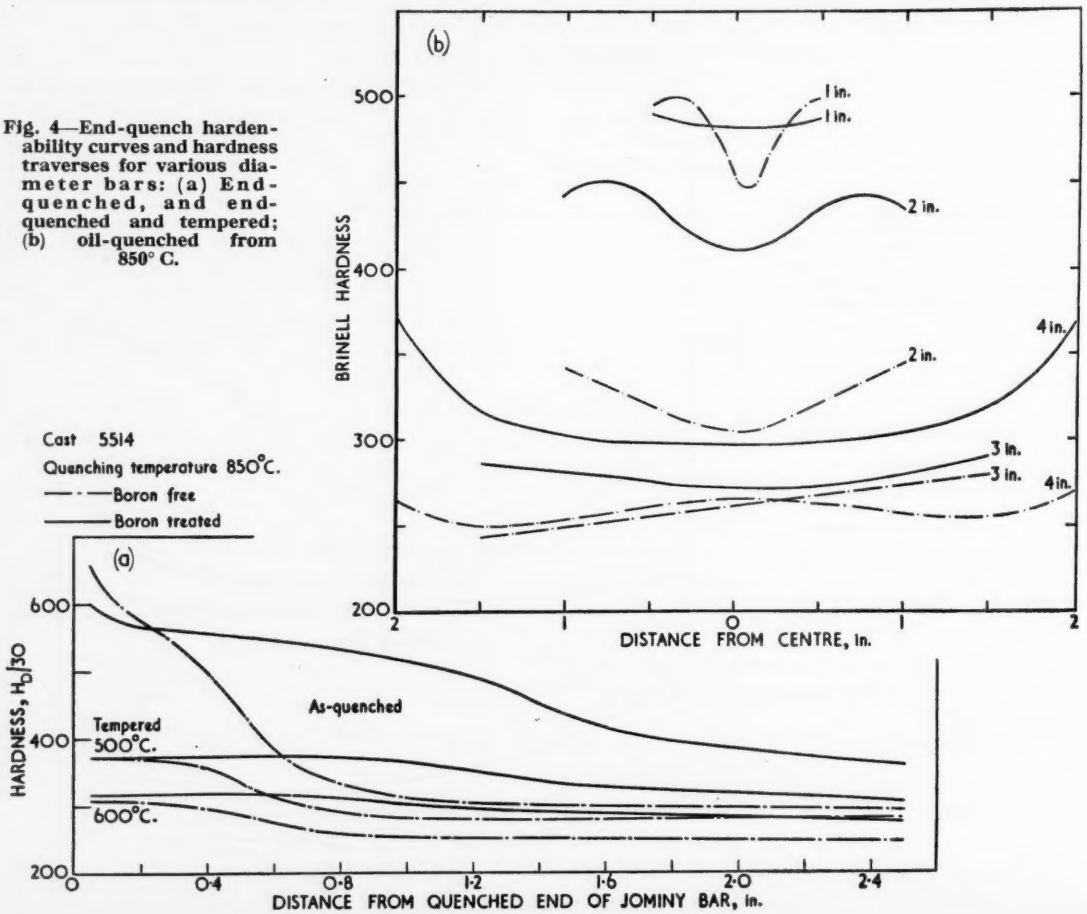


Fig. 4—End-quench hardenability curves and hardness traverses for various diameter bars: (a) End-quenched, and end-quenched and tempered; (b) oil-quenched from 850° C.

Table VIII

IZOD-IMPACT VALUES OF Mn–Mo STEELS FOR EQUAL ULTIMATE STRESS

Effect of Boron, Molybdenum, and Size

Oil-quenched 850° C., tempered 550° C.

Size, in.:			1½		2		2½		4	
Cast	Mo, %	Boron (B) or No Boron (NB)	Ultimate Stress, tons/sq. in.	Izod Impact, ft.lb.	Ultimate Stress, tons/sq. in.	Izod Impact, ft.lb.	Ultimate Stress, tons/sq. in.	Izod Impact, ft.lb.	Ultimate Stress, tons/sq. in.	Izod Impact, ft.lb.
5380	0·03	NB	52·3	77	51·4	81	51·4	72	51·0	71
		B	56·8*	75	53·2	34	50·3	41	48·6	24
5513	0·11	NB	52·9	79	49·4	85	49·9	79	48·0	63
		B	56·4*	71	49·8*	69	51·9	37	49·2	29
5609	0·20	NB	59·2	52	56·0	58	57·6	61	56·0	29
		B	63·7*	66	60·8*	55	57·4*	48	57·8	16
5514	0·28	NB	65·2	59	58·1	69	57·2	69	58·6	44
		B	64·6*	56	60·8*	65	58·1*	63	59·1	28

* Tempered at 600° C.

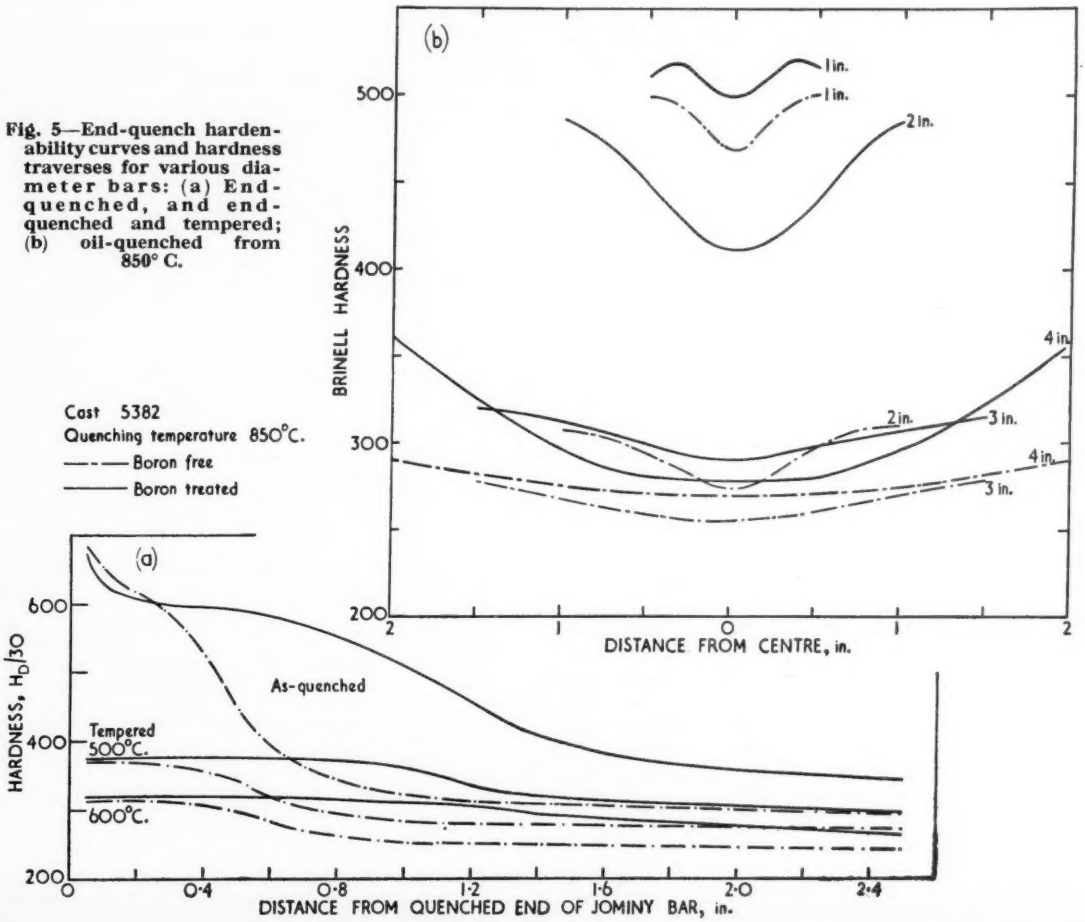


Fig. 5—End-quench hardenability curves and hardness traverses for various diameter bars: (a) End-quenched, and end-quenched and tempered; (b) oil-quenched from 850° C.

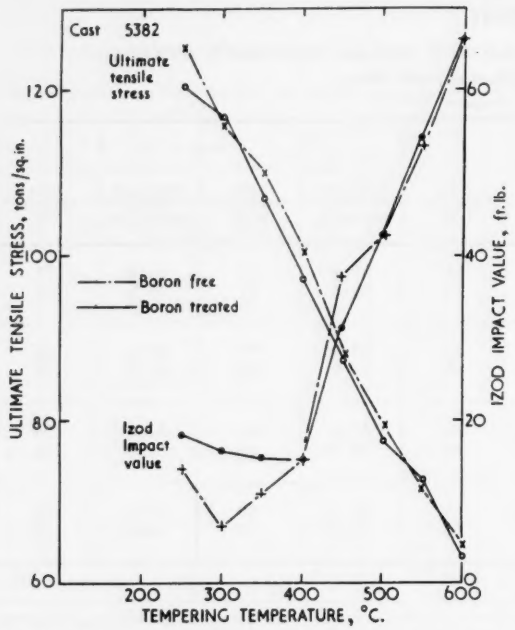


Fig. 6—Effect of tempering on tensile strength and impact value for cast 5382, heat-treated in 0·424 in. dia. and oil-quenched from 850° C.

practical purposes, no difference between boron-free and boron-treated steels (*see* Table VII).

Large Sizes

The size to which an increase in tensile strength with the addition of boron is maintained varies from steel to steel. This can be related to the Jominy hardenability tests, and is also illustrated in Fig. 7, in which the mechanical properties of two of the Mn–Mo series of steels tempered at 550° C. are plotted for various ruling sections. The steel without Mo shows advantage only to a point between 2- and 2½-in. dia., while the steel with 0·28% of Mo retains the advantage to somewhere approaching 4 in. sq. The graphs for the steels with intermediate Mo contents, not reproduced here, show the curves to come together at points corresponding to intermediate sizes. The boron-treated part of the NE 8640 type steel maintains its increase in tensile strength over the non-treated part in all the sizes tested, but this advantage diminishes as the size increases.

The gaps between the curves denoting Izod impact values tend to widen. This is important, for at and above some particular size, varying from steel to steel, the relationship between ultimate stress and impact that holds good for the intermediate sizes no longer applies. Above this point, when the tensile strengths are equal, whether the boron steels have had to be tempered at a higher temperature or not, their impact values are reduced below those of the untreated steels, always significantly and often very drastically.

The ductility values do not show so clear a pattern. Although there is usually a drop as with the impact values, this is not always the case, nor is the magnitude of the deterioration so striking.

The approximate sizes at which this deterioration in properties takes place can be assessed by considering the impact values at equal ultimate stresses, and this is shown in Table VIII for the four steels in the Mn–Mo series. The boron-free steels were all tempered at 550° C., but some of the boron-treated steels (marked with asterisks in Table VIII) had to be tempered at 600° C. to give roughly equal ultimate-stress values.

The sizes at which the impact properties fall off (shown by the thickened line in Table VIII) coincide almost exactly with the sizes at which differential tempering is no longer necessary—that is, where the boron steels are no longer hardening to any appreciably greater extent than the boron-free steels.

Yield Ratio

The yield stresses were determined with dividers, and as on low-alloy steels the yield point is not always well defined, the results recorded must be taken with some reserve. For a few of the tests proof stresses instead of yield stresses were determined, and in these cases the 0·5% offset has been taken as the yield stress. Nevertheless, the information available at tempering temperatures of 500°, 550°, and 600° C. shows certain interesting trends; the data available on lower tempering temperatures are too few to make any reliable inferences.

Judging, therefore, on the high-temperature tempering ranges, in those sizes in which full hardening has occurred, as assessed from the Jominy curves, yield ratios (yield stress expressed as a percentage of ultimate stress) on both boron-free and boron-treated steels are surprisingly high, being mostly in the range 90–95%. On increasing the sizes beyond those in which full hardening was achieved, the yield ratio

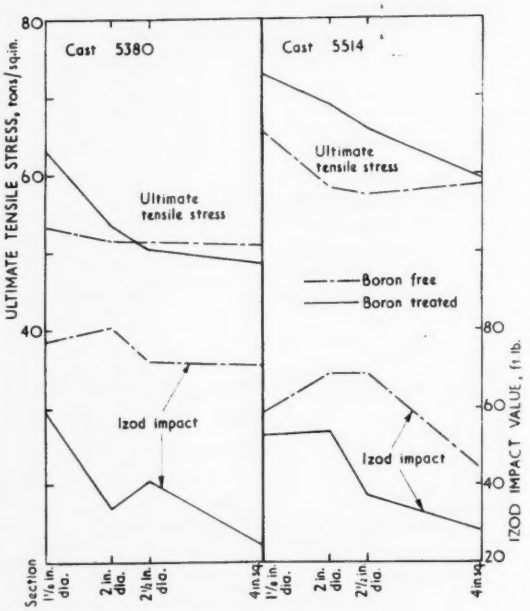


Fig. 7—Effect of section size on tensile strength and impact value for casts 5380 and 5514, oil-quenched from 850° C. and tempered at 550° C.

decreases, sometimes quite suddenly, to lower values, down to about 70%. There is not much difference in this respect between the boron-free and the boron-treated parts of the same cast, except that the size at which the drop occurs is greater for the boron-treated steels (Table IX).

This has had an effect on the mechanical properties of the intermediate sizes. In this range the impact values were essentially similar at equal ultimate stresses for both boron-free and boron-treated steels, although different tempering temperatures were necessary. The boron steels, however, have a marked advantage in yield stress (Table VII). In the smaller sizes, where full hardening occurs in both steels, and in the larger sizes, where it occurs in neither, there is not this advantage.

Notched-Bar Impacts at Sub-Zero Temperatures

Only a few impact tests were carried out on each steel, and as these were in $1\frac{1}{8}$-in. dia. bars, tempered at 550° C., the impact values of the boron steels at 19° C. were generally lower than those for the non-boron steels, and the ultimate stresses were higher. The results show that, with the exception of cast 5609 ($1\frac{1}{2}$% Mn, 0·2% Mo), none of the boron-free steels shows any appreciable drop in impact value on reducing the testing temperature to − 40° C., whereas all the boron-treated steels show a deterioration.

These limited data therefore suggest that the use of boron might have the effect of raising the impact transition temperatures; but it must be noted that in this ruling section the boron-treated steels tested were all fully hardened, whereas the boron-free steels were showing mass effect resulting in incomplete hardening and lower tensile strength.

Jominy End-Quench Hardenability Tests

The Jominy curves in Figs. 1*a*–5*a* show that for all the steels the addition of boron produced a marked increase in hardenability, although very near to the quenched ends the actual hardness achieved is always slightly lower for the boron steels. This increased hardenability is shown in the results of tempered Jominy tests as well as in those in the untempered condition.

The curves have been examined in the light of the tensile and impact test results. A common way of assessing the results is on the basis of 'ideal critical

Table IX

EFFECT OF SIZE AND HARDENABILITY ON YIELD RATIO

Cast	Type of Steel	Boron (B) or No Boron (NB)	Limit of Hardening,* in.	Yield Ratio, %							
				Size, in.							
				0·424	$\frac{3}{4}$	$1\frac{1}{8}$	$1\frac{1}{4}$	$1\frac{3}{4}$	2	$2\frac{1}{2}$	4
Oil-Quenched 850° C., Tempered 500° C.											
5380	$1\frac{1}{2}$% Mn, 0·03% Mo	NB	0·45	93	79	79			75	79	79
		B	1·2	91	96	92			81	73	69
5513	$1\frac{1}{2}$% Mn, 0·11% Mo	NB	0·5			81			82	79	82
		B	1·9			95			87	83	80
5609	$1\frac{1}{2}$% Mn, 0·20% Mo	NB	0·8			89			83	88	84
		B	2·5			96			94	92	90
5514	$1\frac{1}{2}$% Mn, 0·28% Mo	NB	1·05			98			92	88	90
		B	2·7			98			95	93	92
5382	NE 8640	NB	1·05	95		95			92	89	86
		B	2·5	94		98			94	91	88
Oil-Quenched 850° C., Tempered 550° C.											
5380	$1\frac{1}{2}$% Mn, 0·03% Mo	NB	0·45	91	80	75			76	79	71
		B	1·2	91	95	93	86	82	74	70	68
5513	$1\frac{1}{2}$% Mn, 0·11% Mo	NB	0·5			79			79	75	75
		B	1·9			96			87	80	76
5609	$1\frac{1}{2}$% Mn, 0·20% Mo	NB	0·8			83			82	81	84
		B	2·5			95			91	90	87
5514	$1\frac{1}{2}$% Mn, 0·28% Mo	NB	1·05			91			88	86	89
		B	2·7			98			93	93	88
5382	NE 8640	NB	1·05	89		89	84	80	81	81	78
		B	2·5	97		96	92	95	93	92	92
Oil-Quenched 850° C., Tempered 600° C.											
5380	$1\frac{1}{2}$% Mn, 0·03% Mo	NB	0·45	90		76			78	72	72
		B	1·2	92		91			73	69	72
5513	$1\frac{1}{2}$% Mn, 0·11% Mo	NB	0·5			80			79	76	74
		B	1·9			92			84	78	78
5609	$1\frac{1}{2}$% Mn, 0·20% Mo	NB	0·8			85			82	80	77
		B	2·5			92			88	87	80
5514	$1\frac{1}{2}$% Mn, 0·28% Mo	NB	1·05			88			81	82	82
		B	2·7			96			91	87	83
5382	NE 8640	NB	1·05	89		84			80	78	81
		B	2·5	94		95			91	85	83

* These sizes are taken from the Jominy curves; *see* Table X

Table X

LIMITING SIZE TO GIVE FULL HARDENING AS CALCULATED FROM JOMINY CURVES

Cast	Type of Steel	Bar Dia., in.	
		Boron-Free	Boron-Treated
5380	1½% Mn, 0·03% Mo	0·45	1·2
5513	1½% Mn, 0·11% Mo	0·3 (0·5)	1·9
5609	1½% Mn, 0·20% Mo	0·8	2·5
5514	1½% Mn, 0·28% Mo	1·05	2·7
5382	NE 8640	1·05	2·5

diameter,' as calculated by the distance from the quenched end at which a hardness corresponding with 50% martensite is reached.[4] This may be an advantage when dealing with curves showing no marked inflexion, but it seems rather arbitrary. In this investigation, particularly in the case of the boron steels, the Jominy curves as-quenched have this disadvantage, but the change in the tempered curves can be judged more accurately, as there is usually a well-defined flat portion before the hardness begins to fall. The curves tempered at 550° C. have been used to give a rough indication of the sizes in which mass effect becomes operative, and the points at which the curves have fallen 20 D.P.H. numbers from the flat portion have been taken as showing significant falling off in the degree of quench. From these distances on the Jominy bars, the diameters of bars having a similar degree of quench (severity $h = 0{\cdot}6^{-1}$ in.) have been calculated according to the procedure outlined by Russell.[5] These are shown in Table X and are considered to represent the limiting size to give full hardening.

Figure 2*a* shows that the Jominy curve for cast 5513, boron-free, tempered at 500° C., is different from all the other tempered curves; had the diameter been calculated from the curves tempered at 550° or 600° C. it would have been 0·5 in. instead of 0·3 in., and it is likely that 0·5 in. is the truer assessment.

Within the limits of variations to be expected in work of this kind, these figures are in good agreement with the mechanical test results in Tables I–V and dealt with in the discussion of results. For each quality examined, the sections previously designated small sizes are smaller than the diameters quoted for the boron-free steels; the intermediate sizes lie between the two sets of figures; and the large sizes, on which the impact values fall off badly on the boron steels, are larger than the diameters quoted for boron-treated steels.

Based on these very limited data, therefore, it seems that there is a good possibility of being able to use the Jominy hardenability test to indicate the maximum size in which a boron-treated steel can be used without deterioration of impact properties, and to show the range (the intermediate sizes) in which boron treatment can improve mechanical properties.

Hardness Traverses

These curves show very clearly the increase in hardenability achieved, and the effect this has had in certain sizes when the steels are oil-quenched under ordinary laboratory conditions.

ALLOY CONSERVATION

At the inception of this work it was hoped that it might be possible to draw some clear conclusions on the use of boron-treated, lower alloy steels in place of higher alloys, possibly on a basis of hardenability. The data presented, however, make it obvious that no such general conclusions can be drawn if the method of assessment of the steels is to be on the basis of the type of specifications that are used in Britain, requiring certain minima in ductility and notched-bar impact properties, in addition to hardness or ultimate stress as determined in centre or midway positions of the section. With such requirements, size is of the greatest importance. For example, up to 2½-in. dia., boron added to the En 16 steel improves the properties, but in 4 in. sq. its effect is very definitely the reverse, so that whereas the boron-free steel would meet the specification without difficulty, the boron-treated steel barely meets the minimum values specified.

It is suggested, therefore, that any proposed use of boron steels in alloy conservation must be examined critically, with emphasis on ruling section and mechanical properties required for the particular application. If a boron-treated steel can replace a different boron-free alloy steel in a 2½-in. section, it is certainly dangerous to assume it would do the same in 4-in. dia. In an attempt to assess the extent of the savings in alloys that might be achieved, the four steels of the Mn–Mo series, forming a coherent group, have been examined size by size in relation to the specifications En 15, En 16, and En 17.

Cast 5380, 1½% Mn Steel (En 15)

This steel without boron meets B.S.S. En 15 in analysis and mechanical tests, but falls well short of meeting En 16 in ultimate stress in all sizes. From the foregoing discussion, the increase in hardenability due to boron would not be expected to improve its properties in 2½-in. dia., and this is shown to be so in Table XI; further, the boron addition causes the

Table XI

EFFECT OF BORON ON 1½% Mn STEEL

Type of Steel	Boron (B) or No Boron (NB)	Tempering Temp., °C.	Ultimate Stress, tons/sq. in.		
			Size, in.		
			2½	1½	¾
En 15 spec.	...	...	45	...	50
Cast 5380	NB	500	53·1	53·6	55·2*
	NB	550	51·4	52·3	52·7*
	NB	600	48·7	48·6	...
Cast 5380	B	500	53·8	68·5	68·2*
	B	550	50·3	63·3	61·9*
	B	600	49·8	56·8	...
En 16 spec.	...	...	55	60	65

* Tests in ¾-in. dia.

impact value in that size to drop dangerously near to the specified minimum. In 1⅛-in. dia., however, boron enables the steel to meet the En 16 specification, albeit with little margin in ultimate stress, but with adequate impact values.

Cast 5513, 1½% Mn, 0·11% Mo

As with cast 5380, boron enables En 16 tests to be met in 1⅛-in. dia., but not in 2½-in. dia.

Cast 5609, 1½% Mn, 0·20% Mo

With Mo at the bottom limit of En 16 specification, the boron-free steel just meets its mechanical test requirements, although tempering at 600° C. is needed in 4 in. sq. to give the impact values. In 1⅛-in. dia., boron enhances the properties and even raises them to satisfy En 17; but in larger sizes, owing to falling impact values, the steel fails to meet even En 16.

Cast 5514, 1½% Mn, 0·28% Mo

The boron-free steel easily meets En 16 requirements, and in 1⅛-in. dia. can just satisfy En 17 with tempering at 500° C. With boron, En 17 requirements are met in 1⅛-in. dia. more easily with tempering at 550° C., which is an advantage. Boron also improves the properties at 2½-in. dia. so that they just satisfy En 17, but with too slender a margin to be reliable.

Summarizing this examination, an attempt has been made in Table XII to assess the savings in Mo that could be achieved by using boron additions. This assessment, which can only be approximate, varies from quality to quality and depends to a large extent on size.

The use of boron in alloy conservation would therefore demand a more delicate splitting up of specifications into size groups than is usual at present. If this were done on steels without boron treatment some conservation could be effected, although not so much as when boron is used.

CONCLUSIONS

(1) In all the steels tested, boron increased the hardenability as shown by Jominy end-quench tests and hardness traverses.

(2) With steels hardened and tempered between 500° and 600° C. the effect of boron on the mechanical properties varies with size at the time of heat-treatment. Each steel can be divided into three size groups, the limits of which vary from steel to steel:

(*a*) In small sizes, in which both boron-free and boron-treated steels harden fully, boron has no effect

(*b*) In intermediate sizes, in which the boron-free steel does not harden fully but the boron-treated steel does, the boron-free steel has to be tempered at a lower temperature than the boron-treated steel to give equal mechanical properties

(*c*) In large sizes, in which neither the boron-free nor the boron-treated steels harden fully, tensile properties are similar, but the impact values obtainable on the boron steels are inferior—usually very much so—to those on the boron-free steels.

(3) In certain sizes the addition of boron makes it possible to achieve a tensile strength unobtainable in the same steel without boron, except by tempering at temperatures lower than those usually considered suitable.

(4) The possibility of conserving alloying metals, in the compositions studied, by the use of boron when tensile and impact tests are specified is confined to the intermediate size group.

(5) On the basis of specifications En 15, En 16, and En 17, the Mo saved by the use of boron in 1½% Mn steels with Mo varied from nil to about 0·2%, depending on size.

(6) In small sizes tempered at 400° C. or below, boron improves the impact values and, to a less extent, the ductility as indicated by elongation and reduction in area.

(7) Both boron-free and boron-treated steels have higher yield ratios when fully hardened than when partially hardened. Boron imparts no advantage in this respect except that, by virtue of increased hardenability, good yield ratios are achieved in larger sizes than if boron were not present.

Table XII

ESTIMATED MOLYBDENUM SAVINGS BY USING BORON

Note—Where no figures are quoted the addition of boron is harmful

Cast	Type of Steel	Estimated Savings in Mo, %			
		Size, in.			
		4	2½	1⅛	⅞
5380	1½% Mn, 0·03% Mo	...	...	0·2	0·2
5513	1½% Mn, 0·11% Mo	...	...	0·1	0·1
5609	1½% Mn, 0·20% Mo	...	...	0·2	...
5514	1½% Mn, 0·28% Mo	...	0·1	0·2	...

Acknowledgments

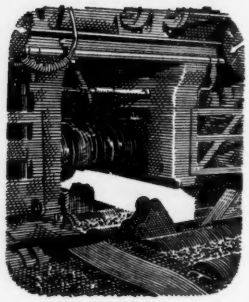
The author thanks the Directors of Samuel Fox and Co., Ltd., for permission to publish this paper, and Mr. G. R. Bolsover for his encouragement and help; and he gratefully acknowledges the contribution of Mr. J. H. Gold. The interest of Mr. F. H. Saniter, Director of Research of The United Steel Companies, Ltd., is greatly appreciated, and for the help in preparing the paper and carrying out some of the tests the author expresses his thanks to Mr. W. E. Bardgett and Mr. F. Gartside of the Research and Development Department.

References

1. W. E. BARDGETT and L. REEVE: *J. Iron Steel Inst.*, 1949, vol. 163, pp. 277–294.
2. Symposium on the Hardenability of Steel, *Spec. Rep. No.* 36., pp. 23–25: 1946, London, The Iron and Steel Institute.
3. *Ibid.*, p. 133.
4. M. A. GROSSMANN: *Trans. Amer. Inst. Min. Met. Eng.*, 1942, vol. 150, pp. 227–255. *Metals Technolog.*, 1942, vol. 9, June, Tech. Publ. No. 1437.
5. T. F. RUSSELL: Symposium on the Hardenability of Steel, *Spec. Rep. No.* 36, pp. 25–33: 1946, London, The Iron and Steel Institute.

IRON AND STEELWORKS ENGINEERING

THE IRON AND STEEL ENGINEERS GROUP



THE IRON AND STEEL ENGINEERS GROUP

REPORT OF THE NINETEENTH MEETING

THE NINETEENTH MEETING OF THE IRON AND STEEL ENGINEERS GROUP of The Iron and Steel Institute was held at 4 Grosvenor Gardens, London, S.W.1, on Thursday, 30th October, 1952. Mr. W. UDALL (Brightside Foundry and Engineering Co., Ltd.) occupied the Chair during the Morning Session, and Mr. R. L. WILLOTT (John Summers and Sons, Ltd.) during the Afternoon Session.

The MORNING SESSION was occupied by the presentation and discussion of a paper on "Electric Drives for Continuous Hot Mills," by Mr. G. R. WILSON of the English Electric Co., Ltd.

The AFTERNOON SESSION was devoted to the presentation and discussion of a paper on "A New Hot Mill for Strips up to 24 in. Wide," by Mr. M. LANGEN of the Loewy Engineering Co., Ltd.

PROCEEDINGS OF THE MORNING SESSION: 10.30 A.M. to 12.45 P.M.

Discussion on the Paper—
ELECTRIC DRIVES FOR CONTINUOUS HOT MILLS*

Mr. G. R. Wilson (English Electric Co., Ltd.) presented his paper.

Mr. G. N. Hewett (McLellan and Partners): There are two advantages of rectifiers that Mr. Wilson has not mentioned. One is that, compared with synchronous motor generators, there is no feed-back into the power system, which may, depending on the circumstances, make large savings in the size and expense of the main switchgear. The other is that a customer buying a rectifier is mainly concerned to have a well-tried tank, so that he will prefer a standard one needing no special design.

The number of phases for larger mills is governed to some extent by the largest tank available. Let us assume, for example, that six-anode tanks are to be used and that 31 tanks are needed to carry the load. Since 31 is a prime number, the possible phases for a balanced system are 6 or 186. A large six-phase system is undesirable because of its low-frequency harmonics, and a system with 186 phases would be costly in transformers. In practice, therefore, either 30 or 32 tanks would be used, and one of the following arrangements would be selected:

30 6-*anode Tanks*		32 6-*anode Tanks*	
Groups of Tanks	Phases	Groups of Tanks	Phases
5 groups of 6	30	4 groups of 8	24
6 groups of 5	36	8 groups of 4	48
10 groups of 3	60	16 groups of 2	96

Mr. Wilson mentions the higher first cost and maintenance of especially close speed-regulation. Mills should, I suggest, be studied in operation to establish what degree of regulation is necessary or economical. There should now have been enough experience of continuous rolling to specify more precisely than in the past what is required of the mill, and so reduce the design of its drive to the real essentials.

There has been a natural tendency in the past to accept U.S. ideas of what equipment is necessary and how large it should be, but it should be remembered that the Americans can afford to be lavish where we cannot.

It has, for example, been discovered by accident that a certain ten-stand mill works just as well with no brakes on the screwdowns; this will save many complications and much money and time on future mills.

Mr. Wilson also dealt with motor powers and converting capacity; in Fig. 7 he showed how an abnormal schedule may be slowed down to bring it within the economical range. A broader picture may be given by plotting the r.m.s. loading of any stand or the whole mill against the tons per hour for each schedule. Figure *A* shows the bounds of a typical nest of curves; the curves end at points representing head-to-tail rolling. Since the commercial output of any mill is limited, probably by the furnaces, to some maximum represented by the

* *J. Iron Steel Inst.*, 1952, vol. 172, pp. 215–225.

vertical line in the figure, any rate of rolling to the right of this line may be ignored; the r.m.s. load need not exceed P in the figure. Need it even be so high? By slowing down the extreme schedule OA, the curve moves to OA^1, and exactly the same output per hour can be obtained at the lower speed and with plant capacity reduced to P^1.

Whether this economy can be made depends on the effect that the slower speed has on the product, and on the importance of that product.

Simpler or smaller plant also means less civil-engineering work. As Mr. Wilson says, using rectifiers instead of motor generators is a step in that direction, since their foundations, housing, and handling are much simpler. In the same way, by designing outdoor-type motors and using mobile cranes, it might even be possible to abolish the expensive motor room.

Any economies, such as those I have outlined, not only reduce first cost and operating and maintenance costs, but also enable the plant to be brought into operation sooner, which is, at present, our most important object.

Mr. **G. R. Wilson**: The two further points quoted by Mr. Hewett in favour of rectifiers are definite advantages, particularly the lack of feed-back into the power system, for the feed-back from large synchronous motors can become a problem.

With the larger rectifier equipments, a greater number of phases are readily obtainable on the equipment. If the transformers are built in 6- or 12-phase units, as would almost certainly be the case, additional phase-shifting transformers would be required in advance of the main transformers to give more than 12 phases on the whole installation.

I agree that design should be cut down to the real essentials wherever possible. This will certainly lower the first cost, and it should also reduce the maintenance required, because there should be fewer individual pieces in the equipment. That also applies to the regulation, but clearly here one must provide what is necessary to meet the requirements of a given mill.

Figure A is very interesting. By slowing the rolling schedule, the motor h.p. load is reduced, but the motor torque load and frame size do not alter. The reduced h.p. loading would, however, reduce the size of the M.G. set, provided that the resulting slower programme is still the extreme programme. The slower programme OA^1 gives the same tons/hr. rate only by reducing the interval time between pieces. In preparing the paper, I had recognized that slowing one programme, as shown in Fig. 7, would automatically give lower output for that product, assuming that there were no other limitations in the mill, such as furnaces. The extra cost of any additional horsepower must also be balanced against the value of any extra output obtained from the mill. Consideration must also be given to the percentage of the total tonnage expected that any programme represents.

The abolition of a motor room and the use of 'outdoor motors' will make the electrical equipment a little more expensive, but the reduction in building costs may produce some overall saving. If this were done, maintenance would be much more difficult, since the motors and control gear would be much more enclosed and maintenance would be weather-dependent. A much more practical solution to the motor-room problem is that now in use in some works, particularly on cold-mill drives, where the motors are enclosed but not necessarily weatherproof and where the control gear is located in proper contactor rooms nearby; everything is still under a roof and so can be maintained in any weather.

I agree that there has been in the past a tendency to copy American steel-mill practice in many ways. This is probably because the American industry is larger than the British, with more equipments of every kind, and therefore with more experience. On the other hand, there are important differences between the British and American industries, and there is no reason why we in this country should not take the lead in improving or simplifying equipment, or in reducing the cost, while still getting the same outputs.

Mr. **P. E. Peck** (British Thomson-Houston Co., Ltd.): There is an essential difference between the regulation required from the motors for a single-strand continuous mill and that for a multi-strand mill. On a single-strand mill, it is best for the regulation of the motors to be equal to the normal I.R. drop.

Curves a, b, and c of Fig. B (i) show the impact and regulation characteristics of three successive motors on a continuous hot mill, percentage speed being plotted against time. The motor regulation is equal to the I.R. drop and the final speeds are reached quickly after the impact swing; thereafter they are coincident.

Figure B (ii) shows the effect of reducing the steady-state regulation by adjusting the machine flux, which inevitably takes time. There is an appreciable area between successive curves before the speeds coincide, which represents looping.

Figure B (iii) shows the effect of correcting the looping of Fig. B (ii) by setting the machine speeds to successively higher values. The presence of metal in the rolls will

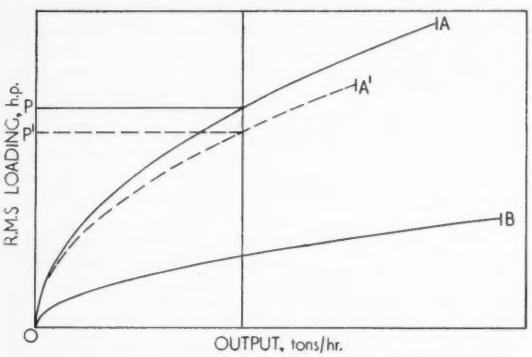


Fig. A—The r.m.s. power load on a continuous mill drive in relation to the hourly output of its various products

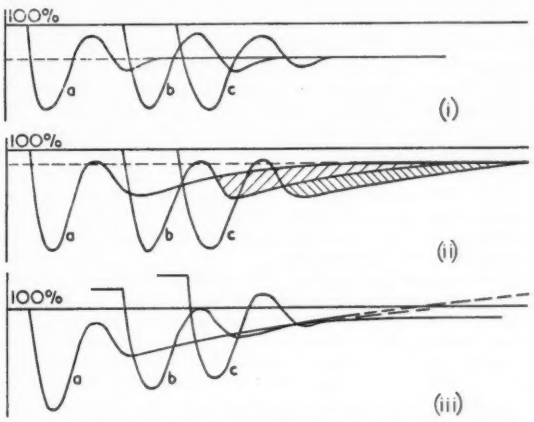


Fig. B—Effects of motor regulation with stands in tandem

prevent the final speeds from being reached, and the result will be tension in the metal and a transfer of load to the later motors.

For a single-strand mill, therefore, motors that reach their final speeds quickly, even if the regulation is large, are better than those in which a close regulation is obtained by a change in the machine flux, or in the flux of some other machine, such as a booster.

On a multi-strand mill, the important factor is that one strand, by entering or leaving, should not disturb the rolling of the other strands. Even so, accuracy can be valued too highly, even in a multi-strand mill. The most difficult case from the regulation point of view is the two-strand mill, which involves a 50% load change when one strand enters or leaves. On a very fast two-strand continuous rod mill installed in this country, the simple compounding exciter has been found to be quite adequate. In this case, the exciter feeds a winding which bucks a main-series compounding winding and so gives a small amount of forcing to the change of flux.

I join Mr. Hewitt in wishing that more information were available about the permitted limits of regulation for the different stages of multi-strand mills. Experience suggests that the limits specified until recently, especially by American designers, were needlessly close and are now being eased.

American mercury-arc rectifier installations are now increasingly of the pumpless water-cooled single-anode or ignitron type, and they are being eagerly adopted for rolling-mill drives, an application in which Europe has hitherto held the lead. Recent discussions with American rectifier designers have indicated that not the least of the advantages of the multi-anode tank is the lower value of faults compared with the ignitron design. This applies both to the rate of rise and to the ultimate value of fault current. It seems to be due chiefly to the greater subdivision of the circuits in the transformer (for the lower power per anode of the multi-anode tank), with consequent higher reactances.

Mr. Wilson's remarks about losses are borne out on one very large installation in this country, where the light-load losses on rectifiers are about one-tenth of those for the equivalent M.G. sets.

The rheostat described on p. 224 of the paper is apparently under hand control; if this is so, and if the one rheostat has sufficient points to give vernier sensitivity, how is that sensitivity obtained on the mechanical side? What steps are taken to avoid the dangerous results of swinging the rheostat from one end of its range to the other?

Mr. **G. R. Wilson**: On the question of regulation, I think the key is differentiation between the two types of rolling. The percentage regulations asked for on the multi-strand mill have sometimes been rather tight. The impact-load application is only a part of the motor's working load on multi-strand rolling; it is one-half on a two-strand mill and only one-quarter on a four-strand mill. The speed changes would therefore not be expected to be due so much to any one strand, so that a motor with a higher regulation than is sometimes asked for may be quite practicable.

In writing the paper, I had in mind the hand-operated rheostat operated by a lever protruding through the desk top. There are several mechanical arrangements that can be adopted. One way of obtaining the necessary sensitivity is to arrange that the full travel of the lever in its slot has a reasonable vernier range. Then, if the operator feels that he is approaching the end of the slot, he can turn the knob on the top of the lever and so cause further movement of the contacts until he feels that he is getting back into range again. Arrangements of this type also avoid the dangerous results of swinging the lever over from one end of the slot to the other.

Mr. **R. Stewartson** (United Steel Companies, Ltd.): Is the diversity factor that Mr. Wilson mentions on p. 222 of his paper the ratio between converting capacity and the total power of the D.C. drive motors, or is it a ratio referring to the converting equipment alone, *e.g.*, in the case of an M.G. set, the ratio between the A.C. drive motor and the D.C. generator? I think it must be the former, and, if this is so, it would seem to depend on the number of stands that the stock passes through at one time, *i.e.*, the maximum demand on the converting equipment. Thus, in a modern rod mill, where the piece is usually in all the stands at once, the factor should be unity, and not tending towards 0·5, as stated on p. 223. As an example, Mr. Phipps* gives the total main-drive horsepower as 6850, and the M.G. set capacity as 5000 kW., *i.e.*, 6700 h.p., in which case the diversity factor is 0·98.

Mr. Wilson says that as much electrical equipment as possible should be installed in a properly constructed and ventilated motor room. This entails an expensive building, with possibly a heavy and little-used crane, near the main mill building. If the main drive motors were mounted on the 'open' mill floor under the mill crane, the adjacent electrical room could be small and light in construction, and, where rectifiers are used for conversion, it might even have no crane at all. Main drive motors are sometimes installed on the mill floor, *e.g.*, in finishing-train drives for rod mills; in a recent installation at one of the branches of my Company, the two main motors of 5500 and 5200 h.p. driving a blooming mill and section mill are disposed in this way with a separate sub-station.

The one drawback in a clean and tidy motor room is the attendant, who may have little to do. During a recent visit to the Manchester–Sheffield–Wath railway electrification, I was impressed by the fact that the numerous sub-stations, over a length of about 60 miles, were unattended, and that all were controlled by a single operator at a central desk. Has such a system any application in a large steelworks with a number of motor rooms and sub-stations? If it were feasible, the initial cost would be high, but, since manpower is becoming increasingly expensive, some economy might result. Such supervisory control would not perhaps pay in an old plant, but would it be an economic proposition in a new plant where, during the early stages of planning, the layout of electrical equipment could be made to suit such control?

Mr. **G. R. Wilson**: My definition of the diversity factor is 'the ratio of the converting kilowatt capacity in the generators to the total equivalent kilowatt rating of the motors on the mill.' I agree that in certain cases the diversity factor on a rod mill might be very near unity, but, in giving the lower figure in the paper, I was trying to indicate the extent of the range; this lower figure is more especially applicable to certain types of rod mill, in which the variety of products (which may not be confined to rods) requires certain motors to be either lightly loaded or switched off when the rest of the mill is running.

I disagree with the derivation of the diversity-factor figure quoted by Mr. Stewartson. The D.C. motors have a total rating of 6850 h.p., requiring an input of about 5400 kW., which gives a diversity factor of 0·93; Mr. Stewartson has, in fact, omitted to take into account the efficiency of the motors. The same source states that

* G. A. Phipps, "Conference on the Design of Rolling Mills," pp. 23–33: 1952, London, British Iron and Steel Research Association.

the synchronous motor rating is 4750 kVA., so that its output is less than 6200 h.p., and this might be as low as 5000 h.p. if the power factor were about 0·8 (a common figure). An M.G. set motor of 6200 or 5000 h.p. is suitable for driving generators with a total output of about 4500 or 3600 kW., which, when taken with the total D.C. motor input of about 5400 kW., leads to the more realistic diversity factors of about 0·83 or 0·67. The presence of three generators totalling 5000 kW. seems to imply that the bus can be split, some motors being switchable to alternative generators, to improve mill flexibility.

Motors might be put into the mill area without much modification or complication, but I hesitate to suggest that protective gear should be similarly treated.

Supervisory control can be made to do almost anything, but it is, however, very expensive. I do not think that it would really pay on a new or an old steel-mill plant because, particularly on the larger plants, there are so many factors involved. It is quite easy to get a system on a large mill which requires 100 indicating lamps on the alarm system. If all those had to be indicated back to a central point, the same amount of equipment might be needed, with additional remote-control switches at the central point. On the other hand, however, valuable time would be wasted in getting an attendant along to the affected motor room to get the mill working again. I think that it is better to have an attendant on the spot or in the vicinity who can look after the equipment immediately, than to have fewer men—probably more skilled—some distance away. Where the motor room may have some of the distribution gear, transformers, etc., in or near it and thus acts as a load centre, there is scope for using a local attendant's time satisfactorily.

Mr. **G. W. R. Fielden** (Metropolitan-Vickers Electric Co., Ltd.): When the mercury-arc rectifier drives are first discussed in the paper, mention is made of using the grid control for starting, and also for compounding after the motor has been brought up to speed and is running on the constant armature voltage. Further power-factor improvement might be achieved if the compounding were done with either on-load tap changing or a virtually stepless Brentford regulator on the rectifier transformer rather than in the grid control circuits.

I agree with Mr. Peck that the current trend in the U.S.A. is towards the pumpless steel tank, instead of the pumped tanks that have been used in the past. Both the Americans and ourselves reckon to get more than five years' life out of these pumpless tanks. American practice, moreover, often uses grid control down to quite low voltages for speed control of mills that are running on load, and not merely for starting purposes.

In the comparison of M.G. sets and rectifiers as a means of conversion, the disadvantage attributed to the rectifiers of having to be highly de-rated when used in high ambient conditions applies more specifically to the general British type of air-cooled rectifier; with the American practice of water-cooling, with thermostatic control, these varying ambient conditions can be combated quite successfully.

I agree that some sort of on-load regulator is necessary on the transformer to overcome the fact that the rectifier is affected by fluctuations in the line A.C. voltage.

It has been implied that rectifiers are not very suitable for motor-room ventilation systems having air going down from the machines to the basement. The water-cooled rectifier would not be subject to this disadvantage; moreover, a suitable baffle arrangement, leading off part of the cooling air, could overcome this disadvantage with up-draught air-cooled tanks.

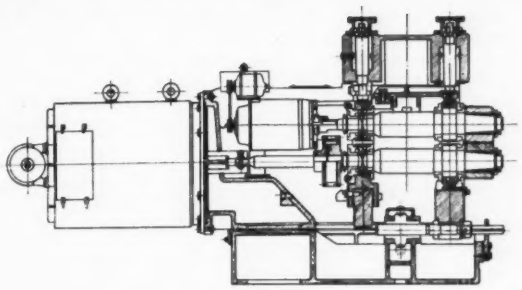


Fig. C—Horizontal rod-mill stand

Mr. **G. R. Wilson**: In answer to Mr. Fielden's first point, a piece enters the average continuous hot mill every minute or minute and a half, and the total load on the mill, particularly on single-strand mills, will change greatly. This may cause a 10% change in the D.C. voltage, which must be taken by the regulator very rapidly. Grid control equipment is rapid, and the maximum effect occurs at light load, falling off with load rise. Its operation is static in that there is nothing mechanical to operate every time. I doubt whether an on-load tap changer would stand up to being swung over a 10% range and back again every minute or minute and a half of its life. Moreover, it would not be quick enough, because each tap change normally takes about 2–3 sec.

The use of a regulating transformer, or some similar device in front of the main transformer, is a possibility, but many of these designs would depend on an on-load tap changer in some way. The grid control equipment is necessary in any case for starting, and so it can easily do the other job in addition.

The American practice of using grid control down to lower voltages on-load may possibly be due to the fact that, in cutting out the tap changer, the equipment is simplified a little. This is possible in the U.S.A., where power systems are larger and more concentrated than in this country.

It is true that the water-cooled rectifier, unlike the air-cooled type, is not so subject to de-rating. The trend is, however, definitely in favour of the latter and away from water-cooling, with its pumps, water next to electrical equipment, etc. On the face of it, having rectifier air coming upwards into a motor room may not seem very suitable for a cool motor-room scheme, but the paper does not go into the full details of what can be done in the way of buildings, baffles, etc., which would obviously be necessary.

Mr. **N. D. Thomson** (United Steel Companies, Ltd.): Mr. Wilson states in the paper (p. 221) that single-anode rectifiers are not favoured in this country for mill drives, although they are in the U.S.A., where advantages in regard to back-fire, robustness, and mounting and replacement are claimed. What are the reasons for this difference in practice ? This matter is of especial interest, since the continuous strip mill described in Mr. Langen's paper* has a drive that uses single-anode rectifiers for the D.C. finishing train.

Figure *C* shows a modern horizontal rod-mill stand using overhung rolls, a type of mill designed in Germany. This mill has driving spindles through the housings, and the main drive is provided by a large motor connected directly to the bottom roll. A small auxiliary motor is connected to the top roll and has the function of merely

* M. Langen, *J. Iron Steel Inst.*, 1952, vol. 172, pp. 203–214.

turning the roll at the same speed as the bottom roll; no pinions are necessary with this design. What is Mr. Wilson's opinion about impact speed-drop in this type of stand? Will the fact that the bottom roll with its motor has a higher inertia than the top roll tend to make the rod sweep up as the front end goes through the stand?

Mr. Wilson seemed to agree with Mr. Stewartson that the diversity factor for a rod mill might be of the order of 0·9 or more, but he states in the paper that diversity factors vary between 0·5 and 0·9, with rod mills tending to the lower figure. Is Mr. Wilson referring to bar mills, and particularly to those where the piece may be in only a few stands at once, when he quotes the lower figure?

Mr. **G. R. Wilson**: Mr. Thomson's first point, on rectifiers, has already been covered; other speakers have mentioned that the trend in the U.S.A. is away from the single-anode tank to the multi-anode tank.

I think there is one mill of the type that Mr. Thomson described, with a low-speed motor, overhung rolls, and an auxiliary motor, but there are as yet little experience and few calculations available for comparison with the more normal type of mill.

Because of this lack of knowledge, it is difficult to say what may happen on impact speed-drop, or whether there may be an upward or downward sweep of the front of the material as it leaves the rolls. Impact-drop arises because a motor has both electrical and mechanical inertia; the impact drops of the main and auxiliary motors can probably be kept the same by careful design, and so the two motors should slow down together and prevent upward or downward sweep. It would, however, seem to be better to have identical motors connected to top and bottom rolls.

The lower diversity factor on rod mills has already been mentioned in the discussion.

Mr. **H. R. Fernbach** (Loewy Engineering Co., Ltd.): Driving the individual motors on continuous mills by single rectifiers, each motor being supplied from one rectifier, would have the advantage of effecting speed control for automatic inter-stand speed-matching on the individual motors by grid control in the rectifier connected into the particular motor circuit. The expense of the individual rectifiers compared with a group rectifier installation may be slightly higher if the rectifier installation is considered by itself, but, on the other hand, the use of grid control for compensating for impact speed-drop of the motors and I.R. drop compensation might simplify the control position. It might be possible to use grid control in conjunction with magnetic amplifiers and to obtain better results than when using electronic amplifiers in conjunction with magnetic amplifiers or regulating exciters in the motor field circuit. With the size of motors used on the mills in question, the single rectifier bulbs would not be unduly small. At present, a large number of individual rectifier bulbs are used in installations with one or two rectifier systems in parallel.

In regard to voltage control of motor speeds by induction regulators or tap-changing switches on the rectifier transformers, at least one installation is now under construction on the Continent in which induction regulators are used for regulating the input voltage to the rectifiers.

Is the high diversity factor of 0·98 on a rod mill due to the fact that the total horsepower of the driving motors has been compared with the horsepower of the motor driving the M.G. set, without taking account of the efficiency of the motors and generators?

Mr. **G. R. Wilson**: Technically there is nothing against having an individual rectifier and grid control equipment for every motor, but it will cost more for the rectifiers alone, owing to replacement of two or three sets of grid control, probably with tap changers, by 15 or 20 sets. There is not yet enough general experience to say whether, in comparison with the alternative methods, it would come out on the total to be less or more expensive. It has probably been worked out for comparison purposes in only a few cases. I think that there would still have to be tap changers or induction regulators, which cost about the same, as well as the grid control for every motor, in order to cover the basic voltage setting on the lines indicated in the paper.

I note that there is now one equipment being made with that form of drive; it will be interesting to see how it works out in practice.

Mr. Fernbach's point about the diversity factor of 0·98 for a particular equipment is partly correct, and has been commented on fully in an earlier reply.

Mr. **J. Nightingale** (Metropolitan-Vickers Electrical Co., Ltd.): The possible development of outdoor electrical installations and drives, which has been mentioned in the discussion, is in my opinion remote. It is doubtful whether the overall economics of such schemes will compare with those of enclosed motor rooms, particularly in view of the machines' greater vulnerability to weather damage during maintenance periods.

Comment has been made on the American development of the single-anode mercury-arc rectifier in contrast to British practice. This development owes its inception largely to the transit damage sustained by the earliest American glass-bulb rectifiers. Development of more robust steel-tank units was initiated and, with a view to using rectifiers on 250-V. D.C. circuits, the single-anode tank, with its promise of lower voltage drop, was adopted. A further point in its favour was its suitability for mass-production manufacturing techniques.

The adoption of individual rectifiers to supply either each pair or successive pairs of finishing-train motors appears to offer the following advantages in addition to those already mentioned:

(i) Reduction in the amount of bus copper and simplification of the bus runs, with attendant cost reductions

(ii) Reduction of short-circuit capacity, with its easing of bus strength and spacing requirements and of the D.C. circuit-breaker rating.

The case for this system will become increasingly valid should horsepowers per stand continue to grow.

Figures 9 and 10 in the paper show graphically the transient response characteristics of two closed-loop regulating systems. For both the full and weak field systems, the better electrical performance is shown by the system of Fig. 9*b*; this has one fewer magnetic lag round the regulating loop than that shown in Fig. 9*a*, which is also burdened with the inclusion of the large magnetic inertia of the mill-motor shunt field. It is noteworthy that the speed-control system that has the better transient response is also the more expensive, largely because of the cost of the booster, which must pass the full armature load current of the mill motor.

Mr. **G. R. Wilson**: Mr. Nightingale's remarks about the overall economics of unattended motor rooms and outdoor installations add weight to my earlier reply.

I believe that his comments about the history of rectifiers in America were correct. There are a large number of 230-V. systems in the U.S.A., where it is a much more common voltage for works supplies than in this country. The lower arc-drop obtainable with the single-anode rectifier is certainly important. The American approach to mass-production of the single-anode rectifier is also understood; the fact that it is single-anode, however, means that there must in any

case be six or twelve times as much production, which increases the scope for mass-production.

The point about a single rectifier per motor in regard to simplification of circuit breakers and copper work is interesting and useful. There is a hint in that direction in the paper where, in connection with circuit-breakers, there is a reference to the advantages of dividing the bus-bar. The logical development of that argument is to have one converting unit per motor. I doubt, however, whether steadily increasing horsepowers per stand are likely. We have already reached the stage on the large hot-strip mill of requiring 5000 h.p. or even more in one motor, and that is not likely to go much further.

The comparison of the curves given in Fig. 10 and the two systems shown in Fig. 9 is the key to the point raised. The armature-circuit time-constant of the machine is usually less than its field-circuit time-constant. Another point in favour of the armature voltage-control method is that there is one fewer time-constant to be considered. But a system that is better electrically probably costs more. It has already been brought out in the discussion that the better electrical system may not always be necessary; each individual case must be considered on its merits.

Mr. **R. L. Willott** (John Summers and Sons Ltd.): A plea has been made for economy in the design of strip-mill equipment. Two wide-strip mills have been in operation in this country since 1939. The overall installation cost of one of them—hot and cold mills, slab-heating and annealing furnaces, pickling and finishing equipment, buildings, foundations, and cranes—was about the same as the present-day price of a fully equipped five-stand tandem mill. Prices have certainly risen since 1939, but the real reason for this low installation cost was stringent economy in design throughout.

It is suggested that hot-mill motors might be outdoor machines, their switchgear and control panels being inside the main building. Economy can be taken too far; a light corrugated-steel lean-to construction utilizing the main building columns will provide all the essential requirements of an efficient motor room at low cost. Anything more elaborate on 'municipal' lines is an unwarranted expense.

It has been stated that we slavishly copy American design. As a user, I think it would be unjust to suggest that British electrical engineering manufacturers in particular cannot meet our precise needs without American technical aid. There is an inquisitive tendency for both user and manufacturer to seek advice in America and to sift current technical literature, but that is not slavish copying of American practice.

We use at Shotton both direct and recirculating air systems, in all cases with heavy maintenance costs. So far we have not tried or investigated electrostatic cleaners; has Mr. Wilson any experience of them ?

Mr. **G. R. Wilson**: Electrostatic precipitation has been used, but, although it is more effective than the other types of filter generally available, it is initially more expensive, and I believe that it takes up more space. One possible way of getting the best of both worlds is to have a recirculating system, with perhaps 10% or 15% of the air extracted and replaced by air drawn into the system through an electrostatic cleaner. That method has been used in this country, and, as far as I know, it is quite successful.

The idea of having the electrical equipment in a room or other construction adjacent to the mill seems to be right. In a few cases it is found that the mill layout requires a wide-span crane, *e.g.*, where cooling beds or finishing equipment come into the mill bay, and in these cases it is sometimes possible for the mill crane to span the motor-room as well. If there is a light removable roof on the motor-room, the mill crane can deal with the rare occasions when electrical equipment has to be moved. This is a practical arrangement and is adopted sometimes, but it is undesirable if its adoption may widen the crane-span still further and so make both the crane and the building heavier.

Mr. **J. L. Gaskell** (Appleby-Frodingham Steel Co.): We ought in this country to give more consideration to supervisory control, either direct wire or coded supervisory. In the comprehensive supervisory scheme at one plant in Utah, I understand that, although all the electrical indications are transmitted back to a central point, there is an attendant in charge of the lubrication of the Ilgner sets. At present I should not like to take the responsibility for an unattended Ilgner set.

The two motors that Mr. Stewartson referred to are single-armature machines with the open commutator away from the mill. They are not open to the external atmosphere, being under the mill roof; four or five years' practice has justified that method of installation for those motors. On the other hand, in another mill, where we have a double-armature machine completely enclosed, a recent occurrence has convinced me that this is not a good arrangement. Wherever the machine is located, the commutator must be fully accessible for running inspections during operation of the mill.

Mr. **G. R. Wilson**: Mr. Gaskell's comments on supervisory control and the desirability of still having an attendant are interesting. Although something can probably be done in this matter, it seems that it will again be a question of economics.

The large motor that is out in the mill and under the mill roof certainly has its open commutator accessible for inspection. What saves the motor from the difficulty of mill dust when it is working is probably the fact that there will be a large air current from the motor ventilating system, which will tend to prevent mill dust from settling in that locality; during a weekend or holiday shutdown, however, when the air system is likely to be off, there is danger of undesirable mill dirt settling on such a motor.

PROCEEDINGS OF THE AFTERNOON SESSION: 2 P.M. to 4 P.M.

Discussion on the Paper—
A NEW HOT MILL FOR STRIPS UP TO 24 in. WIDE*

Mr. **M. Langen** (The Loewy Engineering Co. Ltd.), in introducing his paper, described in more detail the design of the various equipment that was shown in a number of additional photographs. He gave further information on the performance of the horizontal coilers, for which a new design had been developed with a mandrel that can be kept collapsed even at high circumferential speeds. This mandrel is provided with a number

* *J. Iron Steel Inst.*, 1952, vol. 172, pp. 203–214.

Table A

REDUCTIONS FOR 0·049-in. STRIP

Mill	Reductions, %		
	Stand F.4	Stand F.5	Stand F.6
Rehon mill	33·6	30·6	16·9
Typical 2-high narrow-strip mill	21	14	10

Table B

REDUCTIONS FOR 0·099-in. STRIP

Mill	Reductions, %		
	Penultimate Stand	Final Stand	Total Redn. in 2 Stands
Rehon mill	29·6	16·8	41½
Typical wide-strip mill[1]	16–27	5–9	20½–33½
Steel Co. of Wales[2] (0·083-in. strip)	20·75	17	34

of slots in which shoes can slide radially, the outward movement being arrested when the outside diameter is equal to that of the expanded mandrel.

In starting the coiling operation with a collapsed mandrel, the shoes being flung out through the centrifugal force, the leading end of the strip can slide along the second layer after the first lap, as these shoes form a cushion and prevent direct contact, and therefore hammering, of the guide rollers. The mandrel can be expanded after only a few windings, making a forceful contact with the inner layer and providing all the friction that is required for any desired torque applied through the mandrel alone. Thus the guide rollers can be swung out as soon as the mandrel is expanded. In eliminating the hammering of the guide rollers, any marking of the strip is avoided and the guide rollers and other mechanism are not subject to any heavy loads; therefore no undue maintenance is required. This new design has proved to be very successful, enabling the efficient forming of a tight and even coil, without telescoping, in widths with a range of 1–6 in., and in strip gauges from 0·04 to 0·25 in. or thicker.

The average output of the mill has in the meantime been increased to 50 tons/hr. for strip widths above 7½ in. Outputs well above this figure have been achieved when charging hot slabs into the furnace.

Spheroidal graphite cast-iron rolls are now employed in the roughing mill and have proved to be superior to the original cast-steel rolls.

With increased experience in operating the mill, the quality of the products has been further improved, and total differences in strip thickness can be kept within ± 0·003 in.

Mr. **R. Stewartson** (United Steel Companies Ltd.): I should like the views, not only of Mr. Langen but also of the meeting in general, on the desirability of a mill of medium width in this country. My own company operates a narrow-strip mill with 2-high stands. Could Mr. Langen indicate what, in his opinion, is the limit in width for the 2-high arrangement? As shown in Table *A*, the reductions taken in a mill of this type seem to be much less than in the mill described by Mr. Langen.

A comparison between reductions in wide-strip mills is given in Table *B*.

It seems that relatively heavy reductions are being taken in the last stands at Rehon. May I ask Mr. Langen whether this has any adverse effect on surface finish or roll wear (*i.e.*, the frequency of roll changing)? A good surface finish is desirable on cold-rolled strip, which constitutes one quarter of the output.

A rolling speed of 2400 ft./min. is high, and there is the possibility of going to 2800 ft./min. in the course of time. Does this high speed cause noticeable roll wear, and does it give less variation in gauge from the front end of the coil to the back, due to cooling, than would be expected in a wide-strip mill? The newest wide-strip mill in this country has a finishing speed of 2000 ft./min., for which speed "The Modern Strip Mill"[1] quotes 0·004–0·006 in. variation from front to back.

Are the rolls in this mill cambered, and, if so, do they have to be tilted slightly to prevent the strip from becoming wedge-shaped when rolling off-centre? Would such tilting of the rolls have any adverse effect on the bearings, and does the type of bearing—*e.g.*, roller as against fluid film—make any difference in this respect?

I should like to ask the author about the run-out table drive, the rollers of which are individually driven by A.C. variable frequency motors. Can he give some details of their design and his view on why, on the Continent and in this country, satisfactory drives of this type are in operation, whereas in America the D.C.-type seems to be more popular? Although the A.C. drive has the virtue of simplicity, it has a disadvantage in delivering its lowest torque at the minimum speeds, when, presumably, the heaviest strip is being rolled and the highest torque is required.

If the American D.C. applications are for wide-strip mills rolling a proportion of flat plate, frequent reversal might be necessary, and the rollers would have particularly high inertia. Tull[3] gives a conversion efficiency of only 65% for an A.C. drive (the conversion being from A.C. to D.C. and back to A.C.), whereas a single conversion of A.C. to D.C. gave an efficiency of 85%. On a wide-strip mill this made a difference of 4 kWh./ton (*i.e.*, 3d./ton at ¾d./unit). Tull also said that, before the war, the capital cost of the D.C. system in the mill that he described was 30% less than that of the A.C. system, mainly owing to the reduction in building requirements; but today the financial position may have changed.

Mr. **M. Langen**: A wide-strip mill cannot be economically operated if it has to roll strip of less than a third of the maximum width for which it has been designed. A high-output mill for medium-wide strip similar to that described in my paper is capable of rolling a wide range of widths at costs that should be below those for strips slitted from wider widths. Moreover, gauge tolerances are superior, and a hot-rolled edge is preferable in almost all cases. In view of these advantages, and considering the high demand for medium-wide, first-quality strip, I believe that a similar plant is fully justified in this country.

The maximum strip width that, in my opinion, can be satisfactorily finished in a 2-high mill is about 14 in. Considering limits of deflection, rolls suitable for strip above this width would become too large for economic reductions.

With regard to the comparison of reductions as shown in Table *A*, it is impossible for me to judge the rolling schedule given for Mr. Stewartson's Company's narrow-strip mill, without knowing the prevailing conditions. As for the reductions in the Rehon finishing mill, it was originally suggested that heavier drafts should be used in the first three stands, and accordingly smaller drafts in the last stands. The mill is, however, operated with reductions as shown in Fig. 8 of my paper, from which you will note that a gauge of 0·099 in. can be finished

in No. 5 stand. When the rolls in this stand are worn out, No. 6 stand is often brought in, thus maintaining the desired surface finish without roll changing. The finishing of the heavier gauges in five or even four stands has proved to be economical, in view of the wide range of gauges to be produced in each shift. For the widest strips up to a thickness of 0·12 in. all six finishing stands are employed, with smaller reductions in the penultimate stand.

With drafts of 16–17% in the finishing stand, no adverse effects on surface finish and no undue roll wear have been experienced.

Regarding roll wear, the figures given in my paper are, I think, satisfactory. With higher speeds in the finishing mill, the temperature difference between head and tail of the strip is reduced, resulting in closer tolerances. A further reason for employing such high speeds is the increased output of narrow strip.

The work rolls are only cambered when rolling strip above 18 in. wide. When rolling off-centre, cambering is impracticable, but the main screw taking the heavier load has to be readjusted to prevent a wedge-shaped strip. This very slight tilting of the top rolls has no adverse effect on the roller bearings, nor should it have an adverse effect on oil-film bearings if the chocks are properly designed.

The squirrel-cage motors for the runout table incorporate high-resistance rotors, and are of the totally enclosed type. They have a nominal torque of 8·5 lb.ft. and a maximum torque at standstill of 17·5 lb.ft. at 380 V., 50 cycles.

These torque figures are very much smaller than those for motors used on wide-strip mills, and it is my opinion that in this country, and even in the U.S.A., A.C. motors would be very seriously considered for similar applications. The disadvantage of lower torque at reduced speed is not so serious as it seems, because, by adjustment of the alternator voltage, it would be possible to maintain the full torque; and even when the torque is reduced while running at lower speed, it is still adequate and does not practically increase the accelerating time. The strongest point in favour of A.C. motors is their proved reliability and the minimum of maintenance that they require.

I think that the 4 kWh./ton mentioned by Mr. Stewartson represents the total requirements for roller tables and not the saving when using single conversion against double conversion. If D.C. motors were used for runout tables in the installation under discussion, the saving would be only a fraction of a kWh./ton. Conditions have changed since Mr. Tull's paper was published, and today an installation with such small D.C. motors would be more expensive.

Mr. **N. D. Thomson** (United Steel Companies Ltd.): The 24-in. mill has some very interesting and unusual features: for example, the horizontal scale-breaking stand, the flying shears, and the tandem stands in the roughing train.

The blooming mill on this plant has to supply two other strip mills (although I think that one of them will be shut down, as its range is covered by the new mill) and also a rail mill. The output of the mill that Mr. Langen has described will be about 100 tons/hr., and this, added to the blooms and slabs required for the above mills, seems to be rather a large commitment for the elderly blooming mill that is installed in this plant. The mill is operating at present on one or two shifts, but I wonder whether the capacity of the blooming mill will ever allow work on three shifts. In France it is not uncommon for finishing mills to operate on, say, one shift, where a blooming mill has not sufficient capacity to supply all the mills at the same time, but it is difficult to understand how it can be economical to spend all this money on such a continuous mill and operate it for only one or even two shifts. I do not think that we should do it in this country.

The horizontal scalebreaking stand in the mill produces good results and seems to dispose of scale in a much better manner than does a vertical edging stand. Perhaps this accounts in part for the excellent surface quality of the strip. I should like to know Mr. Langen's opinion as to the maximum width of strip for which he would recommend this type of scale breaker. If wider widths (say 30–35 in.) are attempted, will it still be possible to turn the slab up and use a horizontal stand?

Mr. Langen mentioned in the paper that the tolerance obtained in the mill is about ±0·004 in. Just now he gave the tolerance as ±0·003 in.; but the figure given to me by the mill manager was ± 0·002 in. I believe that the post area of the housings is about 280 sq. in., and I wonder how much of the close tolerance obtained is due to this great rigidity. Hessenberg and Sims[4] have pointed out that even if the housings were made infinitely rigid, the rolls would still allow considerable spring. Therefore, is all this great rigidity absolutely necessary? I am thinking, of course, from the point of view of capital cost, where anything that can be done to reduce this will obviously be greatly appreciated by mill users.

The flying shear at the entry to the finishing mill is of admirable design. I have seen a number of flying shears in wide-strip mills, but none with such smooth operation. The torsion bar device is very clever.

On the Continent, fabric bearings are often used, and the roughing train in this mill uses them for the roll necks. I know of one plant that uses fabric bearings on the roll necks on all the mills. They have low friction, low power consumption, and are relatively trouble free and low in cost.

The author has made a very admirable alteration to the coilers to avoid marking on the first few turns of the strip. I should like to know what he considers to be the minimum width for which horizontal coilers can be used on this new design and the maximum width for which vertical coilers can be used.

At a new strip mill recently installed in Sweden by another British firm, a continuous finishing train follows the semi-continuous roughing train. The interesting point is that in the continuous finishing train of 5 stands the first three are 4-highs and the last two are 2-highs. I do not know whether this was the idea of the mill builder or of the user of the plant, but the aim apparently is to obtain a better surface on the strip, due to the burnishing action caused by rolling with a larger diameter. I should like to know whether Mr. Langen thinks that this is worth while.

Another controversial point that often arises is whether it is more economical to roll strip of, say, 30 in. width in a mill specially designed for that width, or to roll it in a bigger mill for, say, 60 in. widths, and to slit it. The fact that certain customers will accept only a rolled edge influences this point, but Mr. Langen has given us figures—based on cost per ton—of power consumption and roll cost in the 24-in. mill. No doubt some people present have figures in mind for the wider mills. I wonder whether we can reach a conclusion on this problem. The manning aspect seems to favour rolling in the wider mill and slitting, whereas the depreciation and cost of slitting favour the narrower mill.

Mr. **M. Langen**: The mill is now working two shifts with satisfactory economic results, which I agree are mainly dependent on the costs for the installation. These costs, however, may not be so great as Mr.

Thomson apparently imagines. An increased output and a three-shift operation will be possible when the present primary mills, which are mainly employed to feed the new mill, have been modernized. The maximum width for a horizontal edging stand is not much above 24 in., because the mill would become too cumbersome for still wider widths and difficulties might arise when entering the rolls. As mentioned earlier, the total tolerance that can now be maintained is ± 0·003 in. Still closer tolerances of 0·002 in., or even below, can be obtained for special orders.

It is our practice to subject housings for such mills to a maximum calculated stress of not more than 1800 lb./sq. in. This makes the stand not absolutely rigid, but keeps the elongation within very small limits, which improve the strip tolerances and therefore justify the weight of the housings.

The minimum strip width for horizontal coilers of the new design is about 3 in., and there are other designs of horizontal coilers available for still narrower strips. A vertical coiler has always the disadvantages of higher labour costs and insufficient temperature control, and will become very inefficient when dealing with strips above a width of 6 in.

I do not think that in this case the strip surface could be improved when using 2-high mills for the two last finishing passes. The Swedish mill mentioned by Mr. Thomson rolls a maximum width of only 12 in., for which 4-high finishing stands are not essential.

The production costs of medium-wide strip rolled in a specially designed high-output mill should be less than those for strip slitted from wider widths, which is also inferior with regard to tolerances and even to surface finish.

Dr. **L. N. Bramley** (B.I.S.R.A.): Apparently the 24-in. mill is running with a gauge tolerance of ± 0·004 in. or less. I should like Mr. Langen to confirm that the possible variation of 0·008 in. is spread along the length of the strip so that it is thin at the front and thick at the back. If the tolerance were improved, would the strip be more valuable and have a better market ?

Is this variation in thickness caused by the fact that when the strip is rolled, it is hotter at the front end than at the back end ? Presumably, the back end cools more than the front end, because it reaches the rolls later and is therefore more stretched in the mill, and this produces a thicker strip at the back end. We have found this factor in one mill in this country, and we have been informed that if this taper could be reduced, it would be a worthwhile improvement.

I noticed in the description of the mill—and I have seen it in several other mills—that the billet is delivered from the furnace sideways; that is, it drops out all at once. If we are right in believing that the taper on the strip is caused by one end being hotter than the other when it is rolled, it might be advantageous to deliver the hot billet out of the furnace in a different manner; for instance, by pushing the hot billet gradually out of the furnace in about 40 seconds, assuming this to be the time required for the rolling operation.

In B.I.S.R.A. we have been studying the problem of the control of gauge of hot strip, and we have under development—and actually just undergoing trials—a very simple and cheap device that we hope will control this variation in thickness along the strip. It is not a precision instrument nor fully automatic, but a pre-set programme control device operating at a very low power level. It can be applied to an ordinary-series motor, and we think that it may have applications other than that of controlling the variation of gauge along the length of hot strip.

Mr. Langen: As already mentioned, maximum tolerances are now ± 0·003 in., and they can be improved for special orders. These tolerances are much smaller than was expected by the production people and, in this respect, they should satisfy all demands of the consumers. The thickness of the strip increases towards the end, owing to the reduced temperature.

In a plant that requires more than one furnace, it is not possible to keep part of the slab in the furnace while it is already entering the roughing mill. The difference in temperature between strip head and tail can be improved by heating the slab end to a higher temperature, as is done in this case, but still better results are achieved with high speeds in the finishing mill, which reduce the waiting time of the strip tail on the delay table.

Mr. **A. Witherington** (I.C.I.): The author and other members have referred to the gauge accuracy obtained from this unit (falling between ± 0·004 in.) over a strip width of 4–24 in., thickness 0·04–0·25 in., and consistent throughout the shift. How is the gauge of hot steel-strip measured and controlled at 2400 ft./min. and with a thickness range of 0·040–0·250 in. ?

The ' over-gauge ' at the cold back end of the strip can be adjusted at speeds up to 500 ft./min. by screwing down on the piece, but the mill operator requires visual indication of the extent of gauge inaccuracy.

Mr. **M. Langen**: The gauge of the strip is measured on every fourth or fifth coil when it has been discharged from the coilers. Such a control is sufficient for maintaining the desired accuracy under normal operating conditions. For checking the characteristics of the mill, accurate measurements across the strip and along its whole length are taken once or twice a week when a coil is rolled out on the floor.

Mr. **S. K. Dean** (B.I.S.R.A.): I believe that in some of the wide-strip mills of this country there is difficulty in getting the strip cool enough to enter the coiler. Sprays are used, but they are not sufficiently effective. I noticed that the temperature in the coiler described here was 800° C. Does that cause any trouble ?

Mr. **M. Langen**: The strip enters the coiler with a temperature of between 700° and 800° C., and such high temperatures have no adverse effect on the performance of the coilers, owing to their new design. To avoid growth of grain, the strip can be sufficiently cooled on the runout table by a spray station having top and bottom water jets. This station, however, is not required when rolling narrow strip with thin gauge.

Mr. **R. B. Sims** (B.I.S.R.A.): In a rolling mill, the distance between the rolls before the strip is entered (the roll setting) is less than the required thickness of the strip. The separating force in the roll gap, which arises from the plastic deformation of the strip, will lift the rolls apart against the elastic restraint of the mill housings, screws, and rolls. The strip thus emerges at a greater thickness than the roll setting, and the following relationship[5] exists between these quantities:

$$F = M(h - S_o)$$

where F is the total load, S_o is the roll setting, and h is the strip thickness. The constant of proportionality, M, which is equivalent to the mill spring, will be the load, in tons, required to stretch the roll gap by one inch if F is measured in tons and h and S_o are measured in inches.

The mill spring is due partly to the elastic distortions in the housings and screws and partly to the deformation in the roll assembly. Experiments have shown that the roll distortion is of the same order of magnitude as that

of the roll housings. In a mill of the dimensions described by Mr. Langen, M may have the value of about 5000 tons/in. and the total roll load may be 1000 tons. The change in the roll gap will then be $F/M = 0{\cdot}200$ in., about half of which will be due to the rolls, and it will be difficult to vary by changes in design because the choice of roll material is restricted and the roll diameters are largely determined by the size of the product rolled. The stretch in the roll gap during a pass can therefore be reduced only by making the housings so massive that their elastic distortion is negligible, or by keeping the roll force sensibly constant.

The author has stated that his new hot mill will produce strip to within $\pm\ 0{\cdot}004$ in. of gauge. This implies either that the roll force in the last stand is kept to within $\pm$ 20 tons (using figures quoted above) during the rolling of a batch of slabs, or that the housings have been made very massive indeed in order to decrease the mill spring. If the mill has been designed with massive housings, perhaps Mr. Langen would comment on the economic aspects of so equipping a mill at a cost of, say, £150–£200 per ton of housing machined, against the added commercial value of strip rolled somewhat closer to gauge. In view of the present trend to put more and more metal into a mill, this point is one of considerable interest and importance.

The practice of rolling narrow material off-centre is also of interest in the study of mill spring. It is clear that one side of the mill will be loaded more heavily than the other side, and will stretch to a greater extent. To obtain a good delivery of strip, the screw of the more heavily loaded side must be lowered to take up this additional spring. If variations occur in the roll load during the rolling, the more heavily loaded side will clearly stretch more than the other side and cause the rolls to tilt and to give a poor delivery of strip. Would Mr. Langen say whether strip with longitudinal curvature is ever produced on this mill when rolling off-centre?

Members may be interested to hear of the development of a new instrument[6] for measuring off-gauge, distinct from the automatic gauge-control devices that are currently under development at the B.I.S.R.A. Rolling Mill Laboratory. Returning to the equation of the mill spring above, it may be written as

$$F/M + S_o = h,$$

so that if quantities proportional to F/M and S_o are added together, the rolled gauge may be measured. Alternatively, if the strip thickness h is written as $h' + \Delta h$, where h' is the desired gauge and Δh is the off-gauge, the equation may be written

$$F/M + S_o - h' = \Delta h.$$

In one form of the instrument that has been tried out on the experimental mill, the load F was measured electrically by a B.I.S.R.A. roll-force meter;[7] S_o was measured by a slidewire on the screws; and h' was obtained from a 3-decade resistance box. This equipment, which could easily be made sufficiently robust for the hot and cold mills in industry, measured strip thickness to $\pm\ 0{\cdot}0005$ in. over a range of 0·01–0·10 in. It would be relatively cheap to install, and cannot be damaged by a cobble.

The principle of descaling is that a light initial pass is given to the material to elongate it and crack or loosen the scale. High pressure water is then used to blow off the scale. When rolling very thick materials with comparatively narrow rolls, it is well known that deformation in the material is relatively heavy near the edges and light at the centre. The scale would be expected to crack off differentially, and in the edge-rolling of slabs it might not be loosened at all at the centre, unless the reduction were very heavy. As edge-rolling is used extensively, one wonders whether descaling depends more upon the pressure of water brought to bear on the surface than upon the deformation that can be imparted by rolls, and, if so, whether it is possible to eliminate or greatly simplify the scale breaker, by using a simple squeeze from manipulators.

This morning Mr. Wilson showed a graph with a single curve, which he stated was the relationship for h.p.hr./ton, or similar unit, against strip thickness for steels over a wide range of temperatures. The h.p.hr./ton data usually published are by no means single curves; they are usually data of such wide scatter that there is generally $\pm$ 50% variation about a mean. When calculating the load and power for a mill by these methods, does the mill designer attempt to place a mean through the data available, or does he resort to the data in the book "The Modern Strip Mill"? If he resorts to "The Modern Strip Mill," does he have the temerity to reduce the power of the mill motors, screwdown motors, etc., below the figure mentioned there? Perhaps Mr. Langen would tell us how he tackles the problem, for there is a tendency always to add to the power used in a previous design but rarely to reduce it. May I recommend to his notice the recent work[8] by B.I.S.R.A. on this subject?

Mr. **M. Langen**: The maximum stress of 1800 lb./sq. in. in the housings limits the elastic distortions to such a degree that their weight is justified. As the machining costs are not affected by increasing the post section of a housing, the additional costs will be less than £100 per ton. It should be borne in mind that the spring is not only caused by elastic distortions in the rolls and the stress in the housing, and that the rigidity of the stand can be improved by substantially over-balancing the top work and back-up rolls, their chocks, and the main screws.

I have never witnessed a longitudinal curvature in the strip when it is rolled off-centre, after the mill setting has been properly adjusted. This indicates that the roll load does not vary much during the pass. The reduction in the horizontal edging mill is about 0·75 in., and in the vertical stands it is roughly equivalent to the thickness of the ingoing section. With such drafts, the scale is sufficiently loosened even at the centre of the slabs, and the efficient design of the nozzles in the high-pressure water system achieves the complete removal of all scale. High-pressure water alone will not guarantee a sufficient descaling, and the use of a manipulator squeezer would be impracticable in this case, owing to the length of the slab and the delay caused by such an operation.

When calculating the load and power for such a mill, we are using the Ekelund formula with a few modifications of our own. It gives, as in this case, practical results. This mill is neither over- nor under-powered. The energy consumption given in my paper represents an average, and it is estimated by dividing the consumption in kWh. by the appropriate output, during a long period.

The **Chairman** (Mr. R. L. Willott): I notice that, in the recuperator under the slab furnace, Mr. Langen uses balanced negative pressure between waste gases and air, and he quotes a preheat figure of 700° C. Does the fan handle air at that temperature satisfactorily, or does it actually run at some lower preheat?

The paper gives water quantities of 2600 gal./min. for roll cooling and 2900 gal./min. for cooling the strip. This total of 5500 gal./min. represents the amount of water used at Shotton on a 56-in. wide mill. As the quoted coiling temperature is as high as 800° C., these water quantities seem to be too great. Perhaps the

figures refer to installed pumping capacity rather than to usage ?

Mr. **M. Langen**: The quoted air temperature of 700° C. has in the meantime been increased to 800° C., and the fan operates quite satisfactorily under such conditions.

The amount of cooling water mentioned in my paper represents the pumping capacity. Although it might be more than the actual requirements, I find that this capacity is in proportion to that installed in most wide-strip mills.

Mr. **M. Dowding** (Davy and United Engineering Co., Ltd.): I would like to hear more from Mr. Langen on the necessity of descaling the slab in the vertical position. He has said that in the mill at Providence the slab is descaled more thoroughly than in a mill where the slab remains horizontal throughout its journey. Is he convinced that the Rehon strip has a better surface finish than that of the wide-strip mills ?

Mr. Willott has already raised the question of cooling the strip after it leaves the finishing stands. Mr. Langen said that comparatively little water is being used at this point, but this may be because surface quality is not yet important. We have looked closely into this question on the wide-strip mills, and found that generally more water than was installed in the first layout is required on the mills when they have been in operation for a time and, in some cases, the original provision has been doubled.

In the paper, the author tells us that the mill was laid out to roll slab of a maximum thickness of 3 in. and that this limitation was made because a larger slab could not be rolled on the existing blooming mill. One therefore assumes that the layout is really ideal for a maximum slab thickness of 3 in. When he introduced the paper, the author said that Providence are now thinking of installing a new blooming mill, and it may be that they would then be able to roll a thicker slab. Will the mill in its present form be able to roll heavier coils from a thicker slab ? I ask this question because there are many strip-mill layouts in which the roughing-mill centres were reduced to save initial capital expenditure, with the unfortunate result that they cannot today meet the demand for heavier coils entailing thicker slabs.

Mr. Langen has raised the difficult question of the cost of installation of automatic gauge-control equipment. Those who are concerned today with developing this promising new invention would be glad to hear from the strip rollers how valuable automatic gauge control would be to them. Is the problem so serious that it justifies considerable capital expenditure, or is it possible to roll to existing tolerances without losing much yield due to off-gauge scrap ?

Mr. **M. Langen**: I am convinced that the descaling in a horizontal edging mill is more efficient because the scale can fall off freely, and I am also convinced that the strip rolled on this mill has a better surface finish than that from wide-strip mills, for reasons which I have given in my paper. The strip leaving the finishing mill can be sufficiently cooled, and there have been no complaints about surface quality, although the mill has now been in operation for one year. The strip temperature can be disregarded so far as the coilers are concerned, as these are suitable for any temperature. I do not know of a hot mill that rolls a coil much heavier than 285 lb./in. of width, as in this case; and even with a new blooming mill, the slab thickness will remain at 3 in., as only blast-furnace gas is available for the reheating furnace.

I agree with Mr. Dowding that the production people should be more definite with their requirements on tolerances, so as to enable mill builders to investigate in detail the merits of additional control equipment.

The **Chairman**: On that score, I think the boot should be on the other foot. If the manufacturers will say that they can produce apparatus to achieve gauge control within certain tolerances, and will undertake to make and sell that apparatus under normal commercial guarantee, we are quite interested.

Mr. **W. P. Snedden** (British Timken Ltd.): I wish that Mr. Langen had been a little more specific in referring to roller bearings as ' tapered roller bearings.' It would be of interest to know how much the ' tapered roller bearings ' are responsible for the accuracy of gauge produced on this mill and for the correction of gauge imparted by the synthetic bearing-equipped roughing stands.

Mr. Langen surely deserves a greater tribute than he has received, in that his mill is able to roll full width at 2400 ft./min. and narrower widths up to 2800 ft./min. These are fast rolling speeds, probably faster than those of any of the American hot-strip mills rolling similar products on roll necks equipped with roller bearings. Also, the bearings on the back-up and work roll necks of this mill are mounted with a loose sliding fit, thereby eliminating the necessity for explosive or hydraulic pumping arrangements to remove them from the roll necks: they are easily moved by manual labour.

It is interesting to know the limitations of loose-fitting bearings on roll necks, as compared with the troublesome tight-fitting bearings that builders on the Continent and in America adopt for high-speed mills. In England we have already had a measure of success with the loose-fitting bearings, and this will probably increase.

The diameter ratio of the back-up and work roll, which is only 2: 1, is low, but apparently it produces the desired results economically.

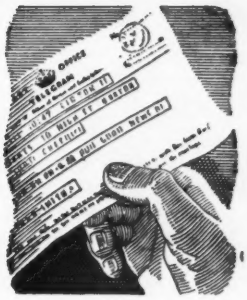
It is difficult to understand how a mill builder can be restricted to a final tolerance on gauge determined by the purchaser of the mill product. Closer tolerances should, of course, be the constant aim of industry, but such tolerances can only be met with improved methods of steel production and precision rolling mills.

Mr. **M. Langen**: I would like to confirm that none of the roller bearings for either back-up rolls or work rolls has so far failed, and that these bearings are partly responsible for the accuracy of gauge. I cannot agree with Mr. Snedden that the synthetic bearings of the roughing mill are causing undue irregularity of gauge. These bearings are giving a very satisfactory performance.

REFERENCES

1. " The Modern Strip Mill," p. 79: 1941, Cleveland, Ohio, Association of Iron and Steel Engineers.
2. H. H. Ascough: *Iron Coal Trades Rev., Special Issue*, 1952, vol. 165, pp. 82–94.
3. I. N. Tull: *Yearly Proc. Assoc. Iron Steel Eng.*, 1949, pp. 662–667.
4. W. C. F. Hessenberg and R. B. Sims: *Proc. Inst. Mech. Eng.*, (A), 1952, vol. 166, p. 76.
5. W. C. F. Hessenberg and R. B. Sims: *Ibid.*, pp. 75–81.
6. R. B. Sims: *Engineering*, 1953, vol. 175, pp. 33–35. The instrument is covered by British Patent No. 27792/51.
7. R. B. Sims, J. A. Place, and A. D. Morley: *Engineering*, 1952, vol. 173, pp. 116–119, 137–139.
8. R. B. Sims: *Proc. Inst. Mech. Eng.*, 1953. (In the press.)

N E W S



ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Special Meeting on Boron in Steel

Wednesday, 29th April

Following the postponement of the Special Meeting which was to have been held on 18th February, 1953, arrangements have been made to discuss papers on Boron in Steel before the commencement of the Annual General Meeting. The programme is as follows:

Morning Session

9.45 A.M.–12.45 P.M.—Presentation of the paper "*Manufacture of Boron-Bearing Steels*," by Dr. H. Rohl (United States Steel Corporation), followed by discussion.

11.0 A.M. (approx.)—Interval for light refreshments

12.45–1.45 P.M.—Buffet Luncheon

Afternoon Session

1.45–4.45 P.M.—Joint Discussion on:

"*The Use of Boron Steels*," by H. B. Knowlton (International Harvester Co.)

"*Effect of Boron on the Mechanical Properties of Low-Alloy Steels*," by R. Wilcock (The United Steel Companies Ltd.) (*Apr.*, 1953).

Annual General Meeting, 1953

The Annual General Meeting of the Institute will be held at the offices of the Institute, 4 Grosvenor Gardens, London, S.W.1, on **Thursday 30th April and Friday 1st May, 1953.** The programme is as follows:

Thursday, 30th April

Morning Session

9.45–10.30 A.M.—Formal business, including:
Report of Council and Accounts for 1952
Presentation of Bessemer Medal
Presentation of other Medals and Prizes

10.30–11.30 A.M.—
Induction of new President
Vote of thanks to retiring President
Presidential Address

11.30–11.45 A.M.—Interval for light refreshments

11.45 A.M.–1.15 P.M.—Joint Discussion on:

"*Train Arrivals, Handling Costs, and the Holding and Storage of Raw Materials*," by M. D. J. Brisby and R. T. Eddison (*Oct.*, 1952)

"*Railway Traffic of the Appleby-Frodingham Steelworks*," by E. R. S. Watkin (*Jan.*, 1953)

1.15–2.30 P.M.—Buffet Luncheon

Afternoon Session

2.30–3.30 P.M.—Discussion on:

"*Developments in the Rolling of Broad Flange Beams at the Cargo Fleet Iron Company, Ltd.*," by G. Barry Thomas (*Feb.*, 1953)

3.30–5.0 P.M.—Joint Discussion on:

"*Speed-Dependent Variables in Cold Strip Rolling*," by R. B. Sims and D. F. Arthur (*Nov.*, 1952)

"*Performance of Hydrodynamically Lubricated Roll-Neck Bearings*," by R. B. Sims (*Dec.*, 1952)

"*Control of Strip Thickness in Cold Rolling by Varying the Applied Tensions*," by R. B. Sims, J. A. Place, and P. R. A. Briggs (*Apr.*, 1953)

"*Works Trial of the 'T' Method of Automatic Gauge Control: A Progress Report*," by R. B. Sims, J. A. Place, and P. R. A. Briggs (*Apr.*, 1953)

Evening

7.0 for 7.30 P.M.—Members' Dinner

Friday, 1st May

Morning Session

10.0–11.0 A.M.—Joint Discussion on:

"*The Quench-Ageing of Iron*," by A. L. Tsou, J. Nutting, and J. W. Menter (*Oct.*, 1952)

"*Effect of Quench Ageing on Strain Ageing in Iron*," by A. H. Cottrell and G. M. Leak (*Nov.*, 1952)

11.0–11.15 A.M.—Interval for light refreshments

11.15 A.M.–1.15 P.M.—Joint Discussion on:

"*Temper-Brittleness: A Critical Review of the Literature*," by B. C. Woodfine (*Mar.*, 1953)

"*Some Aspects of Temper-Brittleness*," by B. C. Woodfine (*Mar.*, 1953)

"*Temper-Brittleness in High-Purity Iron-Base Alloys*," by A. Preece and R. D. Carter (*Apr.*, 1953)

"*Effect of Arsenic and Antimony on Temper-Brittleness*," by G. W. Austin, A. R. Entwisle, and G. C. Smith (*Apr.*, 1953)

1.15–2.30 P.M.—Buffet Luncheon

Afternoon Session

2.30–4.30 P.M.—Joint Discussion on:

"*The Formation of Bainite*," by T. Ko and S. A. Cottrell (*Nov.*, 1952)

"*Effects of High-Temperature Heating on the Isothermal Formation of Bainite*," by S. A. Cottrell and T. Ko (*Mar.*, 1953)

"*Thermal Stabilization of Austenite in Carburizing Steels*," by H. M. Otte and T. Ko (*Jan.*, 1953)

"*Line-Broadening of Martensite in Nickel Steels*," by R. A. Smith (*Feb.*, 1953)

"*The Deformation of Austenite in Relation to the Hardness Characteristics of Steel*," by G. R. Bish and H. O'Neill (*Apr.*, 1953)

Awards

Bessemer Medal

The Bessemer Medal for 1953 has been awarded to Mr. **Richard Mather**, Past-President (Skinningrove Iron Company) for his distinguished services to The Iron and Steel Institute and to the industry.

A portrait and biography of Mr. Mather appear as the frontispiece to this issue.

Sir Robert Hadfield Medal

The Sir Robert Hadfield Medal for 1953 has been awarded to Mr. **W. E. Bardgett**, Research Manager of Central Research Laboratories of The United Steel Companies Ltd., in recognition of his contribution to the development of alloy steels with particular reference to steels for service at elevated temperatures.

Andrew Carnegie Medal

The Andrew Carnegie Silver Medal for 1952 has been awarded to Dr. **E. T. Turkdogan** for his research on slag–metal reactions reported in papers, written in collaboration with Dr. W. R. Maddocks, on "*The Effect of Sodium Oxide Additions to Steelmaking Slags: Parts I and II*" (*Journal*, July 1949 and June 1952) and "*Phase Equilibrium Investigation of the Na_2O–P_2O_5–SiO_2 Ternary System*" (*Journal*, Sept., 1952).

Williams Prize

No award has been made for 1952.

Ablett Prize

The Ablett Prize for 1952 has been awarded jointly (£50 each) to Dr. **S. A. Burke** and Mr. **G. A. Sparham** (British Coal Utilisation Research Association) for their paper on "*Application of Automatic Control to the Coal Feed of Steelworks Gas Producers*" (*Journal*, Nov., 1952).

NEWS OF MEMBERS

▶ Mr. G. P. ACOCK has left the British Iron and Steel Research Association to take up an appointment with the New Zealand Tonnage Committee.

▶ Mr. A. St. H. AUBREY has taken up an appointment with the Sheepbridge Co., Ltd., Chesterfield.

▶ Mr. G. BARBER has left the Steelmaking Division of the British Iron and Steel Research Association to take up an appointment with the Brightside Foundry and Engineering Co., Ltd., Sheffield.

▶ Mr. D. T. BARRITT has been appointed a Joint Managing Director of Simon-Carves Ltd., Stockport.

▶ Professor P. G. BASTIEN, Professor of General Physics and Metal Physics at the Ecole Centrale des Arts et Manufactures, Paris, has succeeded Professor A. PORTEVIN as Director of Scientific Research.

▶ Mr. H. BIGGIN has left the British Iron and Steel Research Association to join the staff of Ashmore, Benson, Pease and Co., London.

▶ Mr. A. L. BRADLEY has left the Steetley Company Ltd., Worksop, to take up an appointment with the Consolidated Rand Brick, Pottery and Lime Co., Transvaal, S. Africa.

▶ Mr. A. R. G. BROWN of the British Iron and Steel Research Association has transferred from the Divisional Laboratory at the Royal School of Mines to the Metallurgy Laboratory, Sheffield.

▶ Mr. D. C. BROWN has left Rotol, Ltd., Gloucester, to take up an appointment with British Messier Ltd., Gloucester.

▶ Mr. E. H. BUCKNALL, Principal Assistant to the Superintendent of the Birmingham Laboratory of the Mond Nickel Company, has accepted the invitation of the Indian Council of Scientific and Industrial Research to take over the post of Director of the recently established National Metallurgical Laboratory at Jamshedpur.

▶ Mr. N. F. DUFTY has left W. D. Scott and Co., Sydney, Australia, to take up the appointment of Lecturer in Metallurgy at Perth Technical College, Perth, Australia.

▶ Mr. C. J. B. FINCHAM has been awarded his Ph.D. degree, and has left the British Iron and Steel Research Association to take up the post of Research Associate at the School of Mines, Columbia University, New York.

▶ Mr. A. G. GILBERTSON, Manager and Director of Brown, Lenox and Co., Ltd., Pontypridd, has been appointed to the Board of the Briton Ferry Steel Co., Ltd.

▶ Mr. R. S. JACKSON has left the David Brown Foundries Company to take up the appointment of Investigator to the Castings Section of the British Non-Ferrous Metals Research Association.

▶ Mr. J. O. LAY of the British Iron and Steel Research Association has been awarded a Fellowship of the Royal Institute of Chemistry.

▶ Mr. B. W. H. MARSDEN has taken up an appointment with the Algoma Steel Corporation, Ltd., Ontario, Canada.

▶ Mr. W. MONTGOMERY has left Frederick Braby and Co., Ltd., Glasgow.

▶ Capt. J. H. NEAD has retired from active connection with the Inland Steel Company, Chicago, U.S.A., and is now residing at Chesterton, Indiana, U.S.A.

▶ Professor H. O'NEILL, President of the Institution of Metallurgists, has been admitted as a permanent member of the General Committee of the British Association for the Advancement of Science.

▶ Professor A. M. PORTEVIN, an Honorary Member of The Iron and Steel Institute, has been awarded the Lavoisier Medal by the Comité des Arts Chimiques of the Société d'Encouragement à l'Industrie Nationale for his development of stainless and heat-resisting steels.

▶ Mr. S. T. QUAASS has left the Ministry of Supply to return to Australia, where he will take up a post in the Aeronautical Research Laboratories of the Department of Supply in Melbourne.

▶ Mr. H. E. N. STONE has left the British Iron and Steel Research Association to take up an appointment at the Fulmer Research Institute, Stoke Poges.

▶ Mr. G. R. WARD has left the Cargo Fleet Iron Co., Ltd., to take up the appointment of Blast-Furnace Manager with John Lysaght Scunthorpe Works, Ltd.

Obituary

Mr. W. FEARNEHOUGH, Chairman and Joint Managing Director of W. Fearnehough, Ltd., Sheffield, on 14th February, 1953.

Mr. T. C. HOGG, of Ayr, on 10th January, 1953.

Mr. D. JEPSON, Head of the Department of Metallurgy, College of Technology, Birmingham, on 24th February, 1953.

Mr. C. LAYCOCK, President of J. Beardshaw and Son, Ltd., Sheffield, on 7th January, 1953.

CONTRIBUTORS TO THE JOURNAL

P. R. A. Briggs, B.Sc., A.Inst.P.—In charge of instrument research and development in the Metal Working Laboratories of B.I.S.R.A.

Mr. Briggs was educated at Normanton Grammar School and at University College, Nottingham. In 1945 he took up a post in the Instruments Section of the Royal Aircraft Establishment, Farnborough, and a year later joined the British Iron and Steel Research Association, where his work has been concerned with instrumentation for research into rolling and for industrial rolling mills.

P. R. A. Briggs

K. Kuo

R. D. Carter, B.Sc., Ph.D.—A member of the Research Organization of British Insulated Callender's Cables, Ltd.

Dr. Carter was born in Cheltenham in 1927, and was educated at Cheltenham Grammar School and at Leeds University, where he was awarded the degree of B.Sc. with first-class honours in Metallurgy in 1947. He continued at the University for a further year, studying the overheating of steel, and then went to King's College, Newcastle-on-Tyne, where for three years he carried out research work on temper-brittleness under the direction of Professor A. Preece. In 1952 he was awarded the Ph.D. degree of Durham University. In August 1951 Dr. Carter joined the staff of the Plessey Company, Ilford, where he worked on the development of nickel-base alloys for high-temperature applications. He took up his present appointment early this year.

Dr. Carter was awarded the Saville Shaw Medal of the Newcastle Section of the Society of Chemical Industry in 1950.

Kehsin Kuo, B.Sc.—Research Chemist, Institute of Chemistry, University of Uppsala, Sweden.

Mr. Kuo was born in Peking in 1923, and was educated at the University of Chekiang, China, where he graduated in 1946. He went to Sweden in 1947 to study metallography at the Royal Institute of Technology, Stockholm. For three years he studied the isothermal transformation of austenite under Professor Axel Hultgren. Since 1951 he has been a Jernkontoret's research fellow at the University of Uppsala, where he has been investigating the problem of carbides of transitional metals under the general guidance of Professor Gunnar Hägg.

J. A. Place, B.Sc., A.Inst.P.—Head of the Rolling Research Section of the British Iron and Steel Research Association, and Deputy Head of the Joint B.I.S.R.A.–Davy and United Engineering Co. Research and Development Team.

Mr. Place was educated at Pudsey Grammar School and at Leeds University, where he graduated with honours in Physics. His studies were interrupted by service with the minesweeping branch of the Royal Navy. He joined the Mechanical Working Division of the British Iron and Steel Research Association in 1949, and has been working on the design of mill instruments and the automatic control of the rolling process.

J. A. Place

A. N. Whiting

Arthur Noel Whiting, L.I.M.—Personal Assistant to the General Works Manager, Appleby-Frodingham Steel Company.

Mr. Whiting was born in 1921, and was educated at Scunthorpe Grammar School and at Scunthorpe Evening Technical School. He joined the Appleby-Frodingham Steel Company as a Metallurgical Inspector in 1939, and three years later was transferred to the Melting Shops as Technical Assistant. He held the post of Assistant Melting Shop Manager from 1947 until 1st January this year, when he took up his present appointment.

Mr. Whiting has been a Licentiate of the Institution of Metallurgists since 1946.

IRON AND STEEL ENGINEERS GROUP

A meeting of Junior Engineers and Operators will be held at Ashorne Hill, Leamington Spa, from Monday to Thursday, 11th–14th May, 1953. Two papers will be presented and discussed: "*Oil Injection Shrink Fitting*," by J. A. Abrahams (English Steel Corporation, Ltd.), and "*Steelworks Waste Heat Boiler Practice*," by R. McDonald (Colvilles Ltd.); and a lecture on "*The Development Plans of the Iron and Steel Industry*" will be given by Dr. T. P. Colclough. The meeting will include visits to and lectures on the Round Oak Steel Works, Ltd., Brierley Hill, and the British Thomson-Houston Co., Ltd., Rugby.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Mr. H. H. MARDON, Head of the Plant Engineering Division of the Association, has relinquished his position to take up the appointment of Chief Project Engineer with Messrs. Ashmore, Benson, Pease and Co. Dr. H. R. MILLS has been appointed acting Head of the Division. He will remain Head of the Division's Mechanical Engineering Section.

Mr. M. W. THRING has been appointed an Assistant Director of Research to the Association. He will continue as Head of the Physics Department.

INSTITUTION OF METALLURGISTS

Examinations, 1953

The next examinations for the Licentiateship and Associateship of the Institution of Metallurgists will be held from 24th August to 1st September, 1953. Candidates must submit their applications for permission to enter the examinations before 1st May, 1953. Application forms may be obtained from the Registrar-Secretary, the Institution of Metallurgists, 4 Grosvenor Gardens, London, S.W.1.

Examinations for the Fellowship will also be held between 24th August and 1st September, 1953, and intending candidates should apply for permission to enter for the examination, submitting, for the approval of the Council, a statement of the subject of the dissertation or the branch of metallurgy in which they offer themselves for examination.

The papers set at the examinations held in 1947–52 are available at 1*s.* per set, and are obtainable on application to the Registrar-Secretary.

NEWS OF SCIENCE AND INDUSTRY

Fifth Liege International Fair

The 5th Liége International Fair will be held from 25th April to 10th May, 1953. It will include metallurgical conferences on the low-shaft furnace and on blowing of the converter, which have been organized by Le Centre National de Recherches Métallurgiques (Section Liégeoise), in collaboration with La Société Française de Métallurgie, Das Verein deutscher Eisenhüttenleute (VDEL), and The Iron and Steel Institute. There will also be visits to steelworks and low-shaft furnace plants.

The Sessions will be held in the Salle Académique of Liége University, Place du XX Août, Liége. Details may be obtained from Monsieur Paul M. Lejeune, Foire International de Liége, 17 boulevard d'Avroy, Liége, Belgium.

Conference on Brittle Fracture in Steel

The West of Scotland Iron and Steel Institute has arranged a "One-Day Conference on Brittle Fracture in Steel," to be held at 39 Elmbank Crescent, Glasgow, on Friday, 15th May, 1953. Papers by authors from America, Sweden, Belgium, and the United Kingdom will be discussed.

Those interested in attending the Conference are invited to send their names and addresses to the Secretary, West of Scotland Iron and Steel Institute, 39 Elmbank Crescent, Glasgow, C.2.

New Arc Furnace at Hadfields, Ltd.

As part of the foundry development plans of Hadfields, Ltd., a 20-ton Birlec Lectromelt arc melting furnace has recently been installed in their East Hecla Works, Sheffield.

The roof may be lifted and swung aside, to expose the whole hearth for rapid charging or fettling, and is suspended at four points from two steel beams, which greatly facilitates roof change.

Pneumatic control is incorporated on the electrode holders, which enables adjustment and replacement to be carried out by remote control. Amplidyne regulation gear is used to provide smooth stepless movement of the electrodes.

Theory of Plastic Deformation of Metals

The H. H. Wills Physical Laboratory and the Department of Adult Education of the University of Bristol, in co-operation with the Institute of Physics, will conduct a short Summer School, followed by a conference on "The Theory of the Plastic Deformation of Metals, with Special Reference to Creep and to Fatigue," in Bristol, on 13th–16th July, 1953.

Further particulars and forms of application can be obtained either from the Director of the Department of Adult Education, The University, Bristol, 8, or from the Secretary of the Institute of Physics, 47 Belgrave Square, London, S.W.1.

Changes of Name and Address

The business assets of the JOHN A. ROEBLING'S SONS COMPANY have been acquired by the John A. Roebling's Sons Corporation, a subsidiary of the Colorado Fuel and Iron Corporation. The John A. Roebling's Sons Company will in future be known as the Roebling Securities Corporation.

The HARLAND ENGINEERING CO., LTD., have opened a Midlands Area Office at Worcester Chambers, 81 Worcester Street, Wolverhampton (Tel. No. Wolverhampton 25482).

DIARY

2nd Apr.—LEEDS METALLURGICAL SOCIETY—"*Recent Advances in Electrodeposition of Metals and Alloys*," by J. W. Cuthbertson—Chemistry Department, The University, Leeds 2, 7.0 P.M.

9th Apr.—LIVERPOOL METALLURGICAL SOCIETY (Joint Meeting with The Iron and Steel Institute)—Annual General Meeting—"*The Basic Processes Involved in the Tempering of Plain Carbon and Low Alloy Steels*," by W. S. Owen—Liverpool Engineering Society, The Temple, Dale St., Liverpool, 7.0 P.M.

10th Apr.—SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS—22nd Biennial Dinner—Royal Victoria Station Hotel, Sheffield, 6.45 P.M.

11th Apr.—SWANSEA AND DISTRICT METALLURGICAL SOCIETY—Annual General Meeting, followed by Presentation of the Prize Winners' Papers—Central Library, Swansea, 6.30 P.M.

13th Apr.—CLEVELAND INSTITUTION OF ENGINEERS—"*Acid Burdening Blast-furnace Performance at Corby*," by A. Stirling—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 P.M.

14th Apr.—SHEFFIELD METALLURGICAL ASSOCIATION—"*Aspects of a Theory of Passivity*," by C. Edeleanu—Grand Hotel, Sheffield, 7.0 P.M.

16th Apr.—STAFFORDSHIRE IRON AND STEEL INSTITUTE—Annual General Meeting—Star and Garter Royal Hotel, Wolverhampton, 7.30 P.M.

20th Apr.—SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS — Film Night — The University Building, St. George's Square, Sheffield, 7.30 P.M.

21st Apr.—AUSTRALASIAN INSTITUTE OF MINING AND METALLURGY—5th Empire Mining and Metallurgical Congress—Melbourne, Australia.

21st Apr.—SHEFFIELD METALLURGICAL ASSOCIATION—"*Modern Rapid Methods of Fuel Analysis*," by R. A. Mott—Grand Hotel, Sheffield, 7.0 P.M.

25th Apr.–10th May—5TH LIÉGE INTERNATIONAL FAIR—Liége, Belgium.

29th Apr.—THE IRON AND STEEL INSTITUTE—Special Meeting on Boron in Steel—4 Grosvenor Gardens, London, S.W.1, 9.45 A.M.

30th Apr.–1st May—THE IRON AND STEEL INSTITUTE—Annual General Meeting—4 Grosvenor Gardens, London, S.W.1, 9.45 A.M.

TRANSLATION SERVICE

(The previous announcement was made in the March, 1953, issue of the *Journal*, p. 296).

TRANSLATIONS AVAILABLE

No. 465 (German). F. RAPATZ and F. MOTALIK: "The Shape of Tool-Life/Cutting-Speed Curves and their Evaluation." (*Stahl und Eisen*, 1952, vol. 72, Dec. 4, pp. 1583–1587).

No. 466 (German). H. BEISSWÄNGER: "The Cupping Test as a Means for Determining the Deep-Drawing Properties of Sheet and Strip." (*Metall*, 1952, vol. 6, Dec., pp. 744–753).

TRANSLATION IN COURSE OF PREPARATION

(German). E. SENFTER: "The Efficiency of Blast-Furnaces and the Way to the Development of the Large Blast-Furnace." (*Stahl und Eisen*, 1952, vol. 72, Dec. 18, pp. 1633–1640).

CHARGES FOR COPIES OF TRANSLATIONS—The charge for translations is £1 for the first copy and 10*s*. for each additional copy of the same translation. Requests should be accompanied by a remittance. These translations are not available on loan from the Joint Library.

ABSTRACTS AND BOOK NOTICES

IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

CONTENTS

	PAGE
Mineral Resources	435
Ores—Mining and Treatment	435
Fuel—Preparation, Properties and Uses	436
Temperature Measurement and Control	436
Refractory Materials	437
Blast-Furnace Practice and Production of Pig Iron	438
Direct Processes	438
Production of Steel	438
Production of Ferro-Alloys	441
Foundry Practice	441
Heating Furnaces and Soaking Pits	443
Heat-Treatment and Heat-Treatment Furnaces	443
Forging, Stamping, Drawing, and Pressing	444
Rolling-Mill Practice	445
Machinery for Iron and Steel Plant	445
Lubricants and Lubrication	445
Welding and Flame-Cutting	446
Machining and Machinability	447
Cleaning and Pickling	448
Protective Coatings	449
Powder Metallurgy	452
Properties and Tests	452
Metallography	462
Corrosion	466
Analysis	468
Historical	469
Economics and Statistics	469
Miscellaneous	469
Book Notices	470
New Publications	472

MINERAL RESOURCES

The Mineral Resources of the Lyonnais. (*Génie Civil*, 1952, **129,** Oct. 15, 388–390). Particular reference is made to the location and form of the St. Etienne coal basin. Iron ore deposits have been worked in the locality in the past, and considerable deposits exist at a great depth, but are not considered worth working.—T. E. D.

Notes on the Iron Ore Deposits in Spain. M. Salis Martinez, L. Barron del Real, J. de Figueras y Figueras, and E. Charro Sanchez. (*Inst. Hierro Acero*, 1952, **5,** Apr.–June, 585–613). [In Spanish].

Llumeres Iron Ores. P. Hernandez Sempelayo. (*Inst. Hierro Acero*, 1952, **5,** Apr.–June, 642–648). [In Spanish]. Geological data relating to Llumeres ores, the most important iron ores of Asturias, are given.—R. S.

Coal, Ore, Salt and Oil Deposits in Central Europe. Hagen. (*Glückauf*, 1952, **88,** Sept., 918–922). The author discusses an extended and improved recent edition of a multicoloured chart of the mineral and ore deposits in Central Europe. The chart, on the scale 1 : 4,000,000, is shown.—P. F.

The Mineral Resources of Turkey and their Utilization for the Home and Foreign Iron and Steel Industries. H. Weigelt. (*Stahl u. Eisen*, 1952, **72,** Oct. 23, 1353–1364). The occurrence and winning of coal, oil, iron, chrome, manganese, and molybdenum ores and refractory materials are described. Developments, past and future, at Karabuk, and the future prospects of the Turkish iron and steel industry are discussed.

Ores and Metals in the Mediterranean Area. (*Metall u. Erz*, 1943, **40,** Jan., No. 1, 15–18 ; Jan., No. 2, 33–36). A survey is given in tabular form of the development from 1913 to 1938 of the chromium, manganese and iron ore industries in the Mediterranean countries. A table shows metal production over the same period.—L. H.

The Ore and Metal Situation of the Far East. R. Dobransky. (*Metall u. Erz*, 1944, **41,** June, 137–141). The ore production and smelting in far Eastern Countries are briefly surveyed, particularly Japan, China, Manchukuo, Korea, and Burma.

The Metallurgist and Metal Supply. J. Boyd. (*Proc. First World Metallurgical Congress*, Sept., 1951, 82–90 ; Amer. Soc. Met., 1952). The author reviews the position of supplies of raw materials for the metallurgical industries. Present conditions are putting a strain on the productive capacities of the Free World. Measures for relief suggested are scientific prospecting, conservation in production and utilization, and anti-corrosion treatment.—D. C. L. P.

ORES—MINING AND TREATMENT

Iron Ores of Swedish Lapland and Their Mining. B. H. Nordlund. (*Jernkontorets Ann.*, 1952, **136,** 7, 175–183). [In Swedish]. The paper deals with proposed measures for increasing iron ore production is Swedish Lapland. Proven reserves of high grade ores are outlined. Plans for reorganizing rail transport and loading facilities in Luleå and Narvik should enable annual shipments of 4–5 million tons from Malmberget and 10–12 million tons from Kiruna to be shipped after 1956.—G. G. K.

Future Iron Ore Mining in Malmberget. T. Ekstam. (*Jernkontorets Ann.*, 1952, **136,** 7, 184–225). [In Swedish]. A detailed account is given of proposed mining methods, drawing and loading of the broken ore, and of its transportation by inclined belts to a central plant for sorting and concentration, involving lifts of up to 200 metres.—G. G. K.

Future Iron Ore Mining in Kiirunavaara. O. Berggren. (*Jernkontorets Ann.*, 1952, **136,** 7, 226–252). [In Swedish]. The proposed transition from open cast to underground mining in this Lapland area is described. The ore will be removed by sub-level caving, and Joy loaders in conjunction with shuttle cars, of 18 tons capacity. On the haulage levels the ore will be drawn by electric locomotives to a central group of eight winding shafts with an annual capacity of 12 million tons.

Iron Ore Development in Quebec and Labrador. W. H. Durrell. (*Eng. J.*, 1952, **35,** Oct., 1053–1057). The author describes the 360-mile railway and airlift needed to develop the Quebec-Labrador iron ore deposits.—J. C. B.

Size Reduction. J. C. Smith. (*Chem. Eng.*, 1952, **59,** Aug., 151–166). Methods of crushing and milling are surveyed and discussed. The selection of size reducing plant in terms of performance and operating cost, and not price, is necessary. American manufacturers are listed. The characteristics and performance of a large range of crushers, rod and ball mills, hammer mills, attrition mills and fluid energy mills are described.—T. E. D.

The Gas Permeability of Fine Materials in the Induced Draught Sintering Process, and Its Determination for Practical Applications. H. Spielhaczek. (*Metall u. Erz*, 1944, **41,** Apr., No. 7–8, 75–19). The necessary physical properties of material for sintering are outlined, with their relation to gas permeability. An apparatus is then described for the determination of the relative resistance to suction of sinter materials, and the possibilities of the application of these results in practice are discussed.—L. H.

The Influence of the Magnetizability of Micaceous Iron Ore on the Roasting of Spathic Ironstone. W. Luyken and L. Heller. (*Metall u. Erz*, 1944, **41**, Apr., No. 7–8, 73–75). It was found that the micaceous iron ore found in small veins in some Slovakian spathic iron ore affected the magnetic properties of the calcined ore. Laboratory experiments indicated that the wasted micaceous ore behaved as crystallization nuclei, and the whole of the ore assumed the hexagonal paramagnetic form.—L. H.

Possibilities of Concentrating Lorraine Ores, and High Intensity Magnetic Separation. L. Coche, D. Dastillon, and L. Wahl. (*Inst. Hierro Acero*, 1952, **5**, May, 476–512). [In Spanish]. The authors give an account of the incomplete French work on the beneficiation and magnetic separation of Lorraine iron ores. Full details are contained in the records at the IRSID research station at Saulnes. The majority of Lorraine ores could not be separated by any physical process under present economic conditions. (42 diagrams)

Beneficiation of Low-Grade Chrome Ore from Dodkanya (Mysore). P. I. A. Narayanan and G. P. Mathur. (*J. Sci. Indust. Res.*, 1952, **11A**, May, 202–204). A good grade chromite assaying 47·5% Cr_2O_3, 23·51% FeO, with 70·6% chromite recovery is obtained by tabling a <100-mesh feed. Electromagnetic separation, Humphrey's spiral, the Wifley table, and selective flotation were unsatisfactory.—E. T. L.

Beneficiation of Low-Grade Soft Chrome Ore from Kittaburu, Singhbum District, Bihar. P. I. A. Narayanan and S. K. Banerjee. (*J. Sci. Indust. Res.*, 1952, **11A**, May, 205–206). Straight tabling of the ore, after grinding it to <65-mesh, and classification into two sizes, produced a combined concentrate assaying, 54·1% Cr_2O_3, 22·4% FeO, with 71·6% chromite recovery. Jigging and tabling of other grades are also reported.—E. T. L.

Beneficiation of Low-Grade Hard Chrome Ore from Kittaburu, Singhbum District, Bihar. P. I. A. Narayanan and M. C. Sen. (*J. Sci. Indust. Res.*, 1952, **11A**, May, 207–209). Straight tabling of the ore after grinding it to <28-mesh, and classification, gave a concentrate assaying 50·72% Cr_2O_3, with 85·5% chromite recovery. Other jigging flotation and tabling arrangements are also reported.—E. T. L.

FUEL—PREPARATION, PROPERTIES, AND USES

New Aspects of the Combustion Mechanism of Solid and Liquid Fuels. S. Traustel. (*B.W.K.*, 1952, **4**, Jan., 17–20). An exposition of D. B. Spalding's papers on the combustion of liquid fuels in gas-streams and the combustion of fuel particles is given. The question of combustion-rate determinations on the basis of data on the shape size, and nature of the fuel, and the type of atmosphere in which combustion takes place, are discussed, and the relations between liquid and solid fuel combustion are explained by means of combustion charts.—P. F.

The Science of Heat and Combustion. H. Schimank. (*B.W.K.*, 1952, **4**, Mar., 73–79). A survey is presented of the historical development of the conceptions and theories relating to the nature of heat and combustion, from earliest times up to the period of Helmholtz amd Joule.—P. F.

Assessment of Fuel, Combustion Air and Combustion Product, with Particular Reference to Fuel Mixtures. A. R. Leye. (*B.W.K.*, 1952, **4**, Sept., 294–299). A method for the assessment of combustion analyses of fuel mixtures is developed, and its application is illustrated by examples.—P. F.

An Ignition-Chamber for Low-Grade Coal. W. Lenz. (*B.W.K.*, 1952, **4**, Sept., 300–302). The development and performance of a radiation chamber for the combustion of low-grade coal dust used for firing boilers is described. The research was initiated by the Mannheim power station and the Kohlenscheidungsgesellschaft.—P. F.

Calculation of Heat Losses by Exhaust Gas by Means of Siegert's Formula. R. Frey. (*B.W.K.*, 1952, **4**, Jan., 20–21). It is shown how Siegert's formula, orginally developed for calculations of sensible heat losses in the combustion of solids, can be adapted for calculations of exhaust heat losses in gaseous combustion, particularly in the cases of blast-furnace gas and producer gas.—P. F.

Measures for Preventing Slag Formation in Steam Boilers. W. Lenz. (*B.W.K.*, 1952, **4**, Apr., 117–119).

Progress in Coal Cleaning. E. Swartzman. (*Trans. Canad. Inst. Min. Met.*, 1952, **55**, 312–317 ; *Canad. Min. Met. Bull.*, 1952, **45**, 518–523). Fine coal-cleaning experiments employing a Hydrotator and Hydrorotator-Classifier combination pilot plant and an Airflow pneumatic classifier are discussed and illustrated. The results are analysed and flow sheets are shown. Coarse coal-cleaning trials using a Stephens Adamson air-sand dry coal cleaner are also mentioned.—T. E. D.

The Accuracy of Sampling Coal. J. Hebden and G. H. Jowett. (*Applied Statistics*, 1952, **1**, Nov., 179–191). The errors involved in the sampling scheme, reduction of sample, and the analysis of an unwashed slack at a colliery were the subject of experiment and statistical study. Improvements were shown to be possible in the accuracy of sample taking and the orders of the various errors are indicated.—D. L. C. P.

Gas from Coals below the Rank of Coking Quality. H. Gratkowski. (*B.W.K.*, 1952, **4**, Aug., 261–262). This is a report of a lecture by G. Traenkner dealing with possibilities of utilizing low-grade coals, including gasification of coal dust with air, with steam and oxygen under pressure, and by the use of solid carriers of heat.—P. F.

Study of the Coking of Lorraine Coals. Langlois and Pamart. (*Institut de Recherches de la Siderurgie Publication*, 1952). A trial, in which 12,000 tons of coke was made from a Lorraine coal blend at the Thionville coke-ovens, is described in detail, giving composition of the blend and coking conditions. The effect of this coke on the operating conditions of a blast furnace (hearth 17 ft. 3 in. in dia.) was studied when used exclusively for 30 days. The results were somewhat inconclusive owing to inadequate measuring facilities on the plant, but a coke rate only 60–80 kg./ton higher than the usual average was considered encouraging.—T. E. D.

Behaviour of Sulphur and Nitrogen Compounds in Gasification Processes, and its Mathematical Evaluation. W. Gumz. (*B.W.K.*, 1952, **4**, Jan., 13–16). It is shown that the combustible sulphur in producer gas and water-gas, in so far as it originates from the combustion process, is present predominantly as hydrogen sulphide with smaller quantities of carbon-oxy-sulphide. Two simple methods for the calculation of both the H_2S and COS contents are developed. An example is worked to show that the H_2S content must not be neglected in calculations, for it exceeds in quantity the amount of methane generated.—P. G.

The Influence of Ash and Clinker on the Operation of Gas Producers. L. G. Townsend. (*Gas World*, 1952, **136**, Nov. 8, 1171–1173). The effect of ash content of coke on blue water-gas manufacture is mentioned. Operating conditions which will help to avoid clinker formation are suggested, and the importance of continuous removal of ash in mechanical producers is emphasized.—T. E. D.

Pressure-Gas Oxyen Coal Gasifier. C. G. Von Fredersdorff and E. J. Pyrcioch. (*American Gas Assoc. Monthly*, 1952, **34**, Sept., 33–35, 46–47). This work was on the oxygen-air, and oxygen-air-steam gasification of pulverized coal at super-atmosphere pressures. The process was proved economically feasible in the United States now that the cost of oxygen has been reduced. Benefits expected for industry include the production of (*a*) a low cost base or carrier gas for blending purposes, (*b*) a gas of intermediate heating value by oil carburation of the hot-make gases from the gasifier, and (*c*) a gas with the H_2/CO ratio adjusted for the synthesis of methane for distribution as pipeline gas.—J. C. B.

On the Calculation of Gasification Equilibria. S. Traustel. (*B.W.K.*, 1952, **4**, Jan., 10–13). Without entering into the question of the attainment of the equilibrium, a simple method for calculating gasification rates is advanced ; it employs a convenient simplification proposed by Deringer.—P. F.

Measuring Instruments and Methods for Dust Determination In Flowing Gases. P. Noss. (*B.W.K.*, 1952, **4**, July, 227–233 ; *Engineers' Digest*, 1952, **13**, Oct., 347–350).

TEMPERATURE MEASUREMENT AND CONTROL

Precision Thermostat for High Temperatures. (*Instrument Practice*, 1952, **6**, Aug., 701–702). The furnace winding acts as the sensitive element, forming one arm of an A.C. bridge circuit, for control of a thyatron tube. Power is supplied to the circuit by a constant voltage transformer ; a variable autotransformer reduces the voltage of a voltage regulator to the value for the required operating range. The bridge output is amplified, with a gain of 6000, to the control grid of the thyatron, which controls the furnace current by changing the impedance of the primary winding of the transformer. A furnace with a temperature range of 1000–1550° C. can be controlled for several hours within ±0·1° C., and for several days to ±1° C.—L. H.

A Furnace Constant-Temperature Controller and Its Modification as a Programme Controller. J. R. Beattie and A. J. Milne. (*J. Sci. Instruments*, 1952, **29**, Sept., 290–292). An instrument of the galvanometer-photocell type is described. Its use as a thermostat is first noted. The thermocouple is Pt/Pt-Rh, and the thermocouple e.m.f. is balanced by a constant e.m.f. from a helical potentiometer. Two photocells, operated from the light lever from the galvanometer mirror, operate relays which vary the furnace winding resistance. For use as a programme controller, the constant e.m.f. source is converted by a synchronous motor and gear train, in such a way that the derived potential equals that of the couple e.m.f. at all points on the required temperature-time characteristic.—J. H.

A Simple Furnace Temperature Controller. M. Shepherdson. (*J. Sci. Instruments*, 1952, **29**, Aug., 263–264). The controller is of the two-position type, with a thermocouple or a D.C. resistance bridge as the receiving element. A mirror galvanometer deflects a spot of light on a photo-cell, which, through a single valve amplifier, controls a relay, operating a mercury switch. The apparatus is simple, comparatively inexpensive and robust, and is reasonably accurate. In one application described, an accuracy of $\pm 4°$ C. has been attained at 1000° C.—L. H.

High Temperature Resistance Thermometer Temperature Controller. W. B. Crandall and P. M. Hackett. (*Bull. Amer. Ceram. Soc.*, 1952, **31**, Oct., 377–379). A low resistance thermometer temperature controller was constructed, consisting of a sensing element, a Wheatstone bridge unit, a galvanometer with capacitor, a Thermocap relay, a regulating reactor, and power contactors. The sensitivity may be adjusted to maintain temperatures from 200° to 1500° C. within ± 1°C.—E. C. S.

Application of Photo-Electric Methods in Industrial Measurement. G. Coeytaux. (*Rev. Gén. Méc.*, 1952, **36**, July, 187–194). The impact of photoelectric and electronic methods on the science of optical mechanics is examined. The principles of photoelectric cells are explained and their application to precision pyrometry is mentioned.—R. S.

Control of Firing Temperatures by Means of Bullers' Rings. (*Ceramics*, 1952, **4**, Sept., 81–84). Bullers' Rings are used in the ceramics industry as a guide to heat treatment at temperatures of 960–1440° C. Details of the grades of rings available are given.—E. C. S.

REFRACTORY MATERIALS

Refractories Policy and Conferences. G. R. Bashforth. (*Refract. J.*, 1952, **28**, Sept., 397–401). This article is based upon the theme introduced by R. Groves dealing with an imaginary conference on refractories. Recommendations are made regarding : (1) Joint consultation committees and the need for careful selection of personnel for these " so as to avoid wastage of time and any interference with discipline within the organization "; (2) training and development ; (3) reference to outside opinion ; (4) economics of refractories ; and (5) the gaining of knowledge.—E. C. S.

Sillimanite in Assam. (*Refract. J.*, 1952, **28**, Sept., 409–411). The properties of sillimanite found and worked in Assam are discussed.—R. A. R.

Further Investigations on Sintering. P. W. Clark, J. H. Cannon, and J. White. (*Refract. J.*, 1952, **28**, Sept., 402–408). The isothermal sintering of the following materials was studied: (1) The pure oxides CaO and Fe_2O_3 ; (2) a commercial dolomite and calcined MgO containing 10% of a eutectic mixture of MgO, CaO, and Al_2O_3 and (3) metallic copper. The average type of packing corresponding to the initial volume porosity (assuming the compacts to consist of uniform, regularly-packed spherical particles), and the modal particle size as indicated by the size distribution curve, were used to calculate n, the number of pores per cubic centimeter of solid material and, from these, the variations in the values of the rate constants.—E. C. S.

Metal Oxide Refractories. (*Ceramics*, 1952, **4**, Oct., 105–112, 119, 128–129). The properties and methods of production of alumina, beryllia, zirconia, zircon, magnesia, thoria, and calcium oxide refractory wares are discussed.—E. C. S.

Some Aspects of the Corrosion of Refractories. N. McCallum and L. R. Barrett. (*Trans. Brit. Cer. Soc.*, 1952, **51**, Nov., 523–548). The dissolution of single crystals of corundum in simple binary slags has been studied, and some measurements of the rates of diffusion in lime-alumina-silica slags under a small concentration gradient have been made. An increase of temperature of 10° C. has a much more pronounced effect on the dissolution rate of alumina than has an increase in the lime from 35 to 55%.—E. C. S.

The Ternary System Na_2O–FeO–SiO_2. P. T. Carter and M. Ibrahim. (*J. Soc. Glass Techn.*, 1952, **36**, June, 142–163). The thermal equilibrium diagram of the FeO–$2Na_2O \cdot SiO_2$–SiO_2 portion of the Na_2O–FeO–SiO_2 system has been constructed from thermal, microscopic and X-ray data. The existence of one ternary compound, $Na_2O \cdot FeO \cdot SiO_2$ has been established. This compound melts incongruently at 976° C., the wüstite formed finally going into solution at 985° C.

Observations on the Conditions of Failure under Load of Refractory Materials. J. Baron. (*Bull. Soc. Française de Ceramique*, 1952, **14**, Jan.–Mar., 3–15, Reprint). Errors in the measurement of the compression of specimens under load are discussed. These are thought to be due to differences between real and measured heating rates and the fact that the whole specimen may not be at a uniform temperature. Some results are shown and improved techniques are suggested to overcome these errors.—T. E. D.

High-Magnesia Periclase Ramming Material. L. W. Austin. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 147–150). This paper deals with a periclase grain of 96–97% MgO formed without the use of fluxing agents, and describes how it is possible to make this material and successfully bond it into furnace bottoms with a decrease of less than 1% in MgO content. The necessary precautions are listed.—G. F.

Use of High-Magnesia Bottom Material. J. A. Voss, jun. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 151–154). The author describes methods of installing high-magnesia bottoms in electric furnaces used for making stainless steels, and discusses their performance.

Refractories for Acid Electric Steel Melting Furnaces. M. W. Demler and J. A. Pierce. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 241–250). The authors discuss the design and construction of the roof, side walls, and hearth of acid electric furnaces, and describe the application of silica, fireclay, and high-alumina refractories. The effects of operating variables on refractory life are outlined.—G. F.

Refractories for Acid Electric Steel Melting Furnaces. C. E. Grigsby, T. M. Mordaunt, and G. S. Kennelley. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 250–257). In this discussion of acid electric furnace refractories, the authors first consider the spalling characteristics of silica bricks and how to prevent this phenomenon. The application of high-alumina and synthetic mullite refractories is then discussed, and finally the design and construction of a furnace are described.—G. F.

Castable Refractories in Electric Furnace Manufacture. T. D. Robson. (*Ceramics*, 1952, **4**, Oct., 134–138). Some uses of " Ciment Fondu " and " Lightning " cements as bonding materials in casting refractories for electric furnaces and electric heaters, are described.—E. C. S.

Ethyl Silicate in Refractories. A. E. Williams. (*Refract. J.*, 1952, **28**, Oct., 453–456). The uses of ethyl silicate as a bonding agent in refractories and in moulds for the lost-wax moulding process are discussed.—E. C. S.

Design and Construction of a Self-Calibrating Dilatometer for High-Temperature Use. A. Lieberman and W. B. Crandall. (*J. Amer. Ceram. Soc.*, 1952, **35**, Nov., 304–308). A dilatometer has been constructed with minimum expense that (1) operates with a specimen $\frac{1}{2}$ to $\frac{1}{4}$ in. high, (2) is self-calibrating, (3) indicates a motion of 2×10^{-6} in. accurately throughout a range of 0·10 in., (4) can be used at temperatures up to 1400° C., and (5) records the data automatically.—E. C. S.

Applications of Thermodynamics in Ceramics. VI. Summary, Bibliography, and Sources of Data. J. F. Wygant and W. D. Kingery. (*Bull. Amer. Ceram. Soc.*, 1952, **31**, Oct., 386–388). The previous papers in this series (May-September issues) are summarized. A bibliography (21 references) and a table of thermodynamic data with a list of sources are also given.—E. C. S.

Methods of Studying the Structure of Molten Silicates. J. O'M. Bockris. (*J. Chim. Phys.*, 1952, **49**, June, c25–c28). Schemes of study fall into the following groups : (1) The identity and measurement of intermolecular forces in the liquid : " degree of dissociation ", and proportion of ionic liaisons. (2) The identity of groups present and the effects of

various ions on these groups. (3) Intramolecular structure; distribution and degree of disorder, and the co-ordination of metallic ions. Methods and theories used in these studies are outlined and discussed.—T. E. D.

Ceramics. F. Singer. (*Indust. Eng. Chem.*, 1952, **44,** Oct., 2296–2308). Developments in ceramics during 1951–52 throughout the world are reviewed. After defining the scope of the subject, the author deals with education, research, analysis and testing, machinery and apparatus, auxiliaries to production, transport and packing, low solubility glazes, silicosis, raw materials, etc. (422 references).—K. E. J.

Comparison of Costs of Insulating Material of Different Thermal Conductivities. W. Niemann. (*B.W.K.*, 1952, **4,** June, 196–197). Nomograms are developed with which the most economic means of heat insulation for tubes can be rapidly determined.—P. F.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

The Stability of Blast-Furnace Shells. M. Esslinger. (*Génie Civil*, 1952, **129,** Oct. 1, 369–372). The idea of making a blast-furnace shell of plates thick enough to be self supporting is explained. The forces to be expected in such a structure are evaluated, and the method of calculation explained. An example of deformation in a model is illustrated.—T. E. D.

Progress in Thermometallurgy. R. Simonet. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 21–25). A survey is made of methods used in extraction and process metallurgy, and the production of iron and other metals in various parts of the world is mentioned.—T. E. D.

Investigations of the Best Design of Hot-Blast Stove. W. Keller. (*Stahl u. Eisen*, 1952, **72,** Oct. 23, 1328–1335). The effect of various methods of constructing Cowper stoves on their efficiency is discussed. Particular attention is paid to the design of the checkerwork and the relationship between heating surface, channel width, weight, and free cross-section per unit volume of packing. Results of experiments on velocity and temperature distribution in the packing are described, and the influence of shape of hood and combustion chamber are considered. The design of burners capable of affording complete combustion in the chamber is briefly mentioned, and a graphical review is given of suggested and existing brick compositions.—J. P.

Blowing Engine Selection. G. A. Flint and J. G. Lee. (*Blast Furn. Steel Plant*, 1952, **40,** Sept., 1070–1075; Oct., 1210–1211). The matching of blowing engines to blast-furnace requirements is considered. Larger blowers than are necessary for current demand are frequently installed to allow for future plant expansion, and these operate at part load and less than maximum efficiency with correspondingly increased fuel costs. Two plants are considered. An analysis of blower operation over periods of 3400 hr. and 3267 hr., respectively, are made. The capital costs, fuel costs, and fixed annual costs of various sizes of blower are considered. Considerable economic savings could have been made if smaller blowers had been installed at first and scrapped when extensions were eventually made.—B. G. B.

Nomograms for Iron Ore Reduction. J. A. Boneda Sopena. (*Inst. Hierro Acero*, 1952, **5,** Apr.–June, 649–659). [In Spanish]. In this mathematical paper, combustion and the reduction of iron ores are studied and a 'reduction nomogram' is constructed.—R. A. R.

Determination of Heat Losses from Blast Furnaces by Measurements on the Cooling Water. A. Send and F. Strähuber. (*Stahl u. Eisen*, 1952, **72,** Nov. 20, 1509–1512: *Iron Steel Inst. Translation Series No.* 462, 1953). A knowledge of heat losses is necessary in making up the heat balance of a blast furnace. The heat losses through cooling water have therefore been determined on three blast furnaces of different ages and shaft lining thicknesses. The same heat losses were found in the furnace bottoms but different losses in the shafts. The total losses varied between 4 and 5 $\times$ 10^6 kcals. per hour and corresponded, at a hearth loading of 1000 kg. per sq. m./hr. to 150–200 kg.cal./kg. of coke. The following expressions are proposed as being of most importance: (1) Heat loss per hour of operation, (2) loss per sq. m. of furnace surface per hour, and (3) loss per kg. of coke. The influence of hearth loading on the magnitude of heat losses was determined and it is recommended that, in reporting for comparison purposes, the hearth loading should be stated or the losses referred to a standard value of hearth loading.—J. P.

Graphical Presentation of Blast-Furnace Operations. J. Wscieklica. (*Bol. Assoc. Brasil. Metais*, 1952, **8,** Jan., 20–31). [In Portuguese]. The author presents a graphical solution to the main equations for blast-furnace processes and studies methods of charcoal conservation. A charcoal consumption of 460–470 kg./ton of iron is achieved in four small blast-furnaces. The pig iron contained 4% C, 0·4% Si, 0·8% Mn, and 0·2% P. Up to 100% sinter made from Jacutinga and Itabirite ores was charged.—R. S.

Modern Experience in the Beneficiation of Iron Ore in the Humboldt Low-Shaft Furnace. E. Killing. (*Inst. Hierro Acero*, 1952, **5,** Apr.–June, 575–584). [In Spanish]. This is a Spanish version of a paper which appeared in *Stahl u. Eisen*, 1952, **72,** July 31, 925–928 (see *J. Iron Steel Inst.*, 1953, **173,** Jan., 97).

Gas Sampling Apparatus for the Control of Blast-Furnace Operation. W. Loorz. (*Stahl u. Eisen*, 1952, **72,** Oct. 23, 1325–1328). It is advantageous, for many reasons, to have a continuous record of the gas flow and the top gas analysis in a blast-furnace. The development of this kind of measurement is discussed and a new gas probe and measuring device are described for taking temperatures and gas samples for analysis. The importance of the ratio of burden volume to coke volume and of the throat diameter to distributor bell diameter is indicated.—J. P.

Smelting in Czechoslovakia. V. Tafel. (*Metall u. Erz*, 1944, **41,** Oct.–Dec., 229–235). An historical survey is first made of metal production in Czechoslovakia, followed by a description of the wartime position between 1939 and 1944. The article deals almost entirely with non-ferrous smelting, but reference is made to the smelting of iron ores and manganese ores, in the former Hungarian territory.—L. H.

DIRECT PROCESSES

Production and Use of Sponge Iron for Steel in Sweden. D. J. Girardi. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9,** 25–38). The author describes the Höganäs and Wiberg-Söderfors processes used in Sweden for the production of sponge iron, and discusses their advantages and disadvantages. The use of sponge iron in steelmaking is also described.—G. F.

Smelting Kuji Iron Sand by the Krupp-Renn Process. K. Kuwada. (*Proc. First World Metallurgical Congress.* Sept., 1951, 188–198; *Amer. Soc. Met.*, 1952). The author describes Japanese iron sands, and shows how the Krupp-Renn process is used to produce sponge-iron from sand from the Kuji mine. Properties of products made from the sponge are described.

PRODUCTION OF STEEL

Evolution and Recent Trends in Iron and Steel Producing Methods. (*Usine Nouvelle*, 1952, Oct. 16, 23–27; Oct. 23, 30–31). Details of efforts to increase production in the iron and steel industries are given by experts from France, Germany, Switzerland, Canada, Sweden, and Belgium, each author mentioning particular aspects of developments in his own country.—T. E. D.

The Austrian Iron and Steel Industry, and Its Latest Developments in the Technical and Metallurgical Field. F. Leitner. (*Inst. Hierro Acero*, 1952, **5,** July, 697–707). [In Spanish]. Production figures for 1950 are given. Details are furnished of Styrian ores and iron production at Donawitz and Linz. The author surveys methods of production in open-hearth and electric plants and describes the distribution of the manufacture of rolled products. Amongst new developments mentioned are the low-shaft electric furnace, use of oxygen, and continuous casting. The Austrians have produced billets and slabs up to 450 sq. cm. cross-section by continuous casting at the rate of 100 kg./min. in each cooled mould. No technical details are given.—R. S.

Could the Past Glories of Converters Return? O. Hájíček. (*Hutnik* (Prague), 1952, **2,** 10, 220–225). [In Czech]. The principles, advantages and disadvantages of converter practice, past and present, are discussed. It is concluded that considerable possibilities lie in the development of converter practice using oxygen and steam. This necessitates automatic control of the blow facilitated by the use of both photo-cells and spectroscopes. Some suggestions for suitable arrangements of this type are made.—P. F.

Metallurgy of the Bessemer Process. H. B. Emerick. (*Blast Furn. Steel Plant*, 1952, **40,** Oct. 1177–1183, 1206, 1212–1213, 1218). A fairly detailed account of modern Bessemer

practice is presented. A description of the converter is given, together with details of working the blow.—B. G. B.

The Problem of Refining Steel with Pure Oxygen. O. Cuscoleca. (*Stahl u. Eisen*, 1952, **72**, Aug. 19, 989–992). The problem of increasing steel production in Austria can be solved only by a process which uses no bought-in scrap and which does not draw on the scarce reserves of producer-gas coal. The process must also produce a material at least as good as open-hearth steel without requiring increased coke consumption in the blast-furnace. Exploratory production over the past two years in a 5-ton and a 15-ton converter has proved the technical and economic feasibility of refining with pure oxygen.—J. P.

Operation and Metallurgy of Steel Refining with Pure Oxygen in a Top-Blown Converter. H. Trenkler. (*Stahl u. Eisen*, 1952, **72**, Aug. 14, 992–997). The Linz-Donawitz (LD) steel-refining process employs pure oxygen blown on to the surface of the hot metal in a basic converter through a water-cooled copper nozzle. It permits the production of a variety of steels with optimum contents of phosphorus, sulphur, nitrogen, and oxygen, this being of great importance for the segregation behaviour in the ingot. All the variables in the refining process have been clarified and it is therefore possible to ensure any desired steel and slag compositions. The iron contents of the slags are normal ; low phosphorus values in the steel are not obtained at the expense of high iron losses in the slag. The refining process produces a deoxidized steel since the phosphorus content reaches acceptably low values before the carbon has been removed to low levels. The favourable thermal efficiency of the process gives great latitude in the choice of hot metal composition and permits this to be adjusted to steelworks' requirements or, alternatively, allows the smelting of iron with the lowest coke consumption.

Plant, Operation and Economic Efficiency of Steelworks Using Oxygen. K. Rösner. (*Stahl u. Eisen*, 1952, **72**, Aug. 14, 997–1004). Operational equipment for the new process of refining steel by top-blowing with pure oxygen has been developed and the plants now under construction at Linz and Donawitz are described. The lay-out at Linz conforms to the usual Bessemer construction, whereas that at Donawitz is of new design which will permit the eventual setting up of converters and open-hearth furnaces on one stage. Both works are designed for three 25-ton converters, of which two will soon be erected. The works will each produce 150,000 tons of steel per year. Linz will have a 1000-ton and Donawitz two 500-ton mixers. The oxygen plant will have an hourly output of 2000 cu. m. of 98–99·5% oxygen. The course of a melt is described on the basis of experience with pilot-scale oxygen plants at Linz and Donawitz and the bases for cost evaluation worked out. Comparison of costs for refining charges, operation, and construction shows that oxygen-blown converter steelworks are more economic than open-hearth steel-works.—J. P.

Use of Oxygen in the Bessemer Converter. W. T. Rogers and L. T. Sanchez. (*Trans. Amer. Inst. Min. Met. Eng.*, **194**, *J. Met.*, 1952, **4**, Sept., 933–938). This paper presents the results of a production experiment carried out at the Lorain works of the United States Steel Co. to evaluate the effect of the use of oxygen in the Bessemer converter. The experience gained has enabled a standard practice using oxygen to be established, which permits considerably more steel scrap and up to 14% of cold pig to be melted. A more uniform production rate can be maintained and the use of oxygen does not appear to affect bottom life, final steel quality, or teeming temperature.—G. F.

Improving Ordinarily Blown Converter Steel. E. Maurer and W. Gilde. (*Met. u. Giessereitechn.*, 1951, **1**, Dec., 354–357). Attempts to remove nitrogen from converter steel by titanium additions were unsuccessful, but it was found that the main differences in behaviour between converter steel and open-hearth steel were due more to differences in phosphorus content than to differences in nitrogen content. Reducing the permissible phosphorus content from 0·090% to 0·060% gave a significent improvement. The cold brittleness of the converter steel disappeared and it was easily weldable.

Operation and Scope of Dolomite Sprays for Fettling Open-Hearth Furnaces. E. Kupfer. (*Met. u. Giessereitechn.*, 1951, **1**, Dec., 374–375). Tests recently made at the Stahl- und Walzwerk Brandenburg with the Riesa and the Gröditz dolomite-spraying machines led to the conclusion that neither machine produced a real spray, the function of the compressed air being merely to project the dolomite across the furnace in a similar way to hand fettling. It is suggested that it would be worth developing a machine which produced a real spray, using dolomite of suitable grain size, from a water-cooled nozzle which could be inserted into the furnace and could be directed against any part of the hearth or even the roof. It should be suitable for use by one man.—H. R. M.

Progress in the Basic Lining of Open-Hearth Furnaces. L. Hütter. (*Stahl u. Eisen*, 1952, **72**, Oct. 9, 1285–1293 : *Iron Steel Inst. Translation Series No.* 461, 1953). Ways and means of improving furnace life are discussed, particular attention being paid to the performance of the separate parts of the furnace. Although it cannot be claimed that a complete answer has been found to all problems concerning all-basic construction, important progress has been made in recent years.—J. P.

A Proposed Steel Making Process. A. Reggiori. (*Proc. First World Metallurgical Congress*, Sept., 1951, 256–259 ; Amer. Soc. Met., 1952). It is suggested that many advantages would accrue from having a special furnace for melting scrap, and charging only molten scrap and iron to the open-hearth furnace. It is estimated that an increase in production of at least 25% could be obtained.—D. L. C. P.

The Use of Combustion Models in Open-Hearth Furnace Studies. A. H. Leckie. (*Proc. First World Metallurgical Congress*, Sept., 1951, 206–220. Amer. Soc. Met., 1952). An account of the results obtained on the open-hearth combustion models set up and operated by B.I.S.R.A. is given. These results pertain to the effects of air/fuel ratios, furnace pressures, and port design on heat transfer to be expected in an open-hearth furnace. The principles of working models are discussed.—D. L. C. P.

Operation of a Burner-Type Open-Hearth. I. Murata. (*Proc. First World Metallurgical Congress*, Sept., 1951, 235–240 : Amer. Soc. Met., 1952). The author describes how a furnace which had been built for burning mixed gas, was converted to use cold coke-oven gas, with fuel oil or tar as illuminant. The results of the first three months' operation are described.—D. L. C. P.

Cockerill Company's Experience with the Perrin Process. J. Janvier, M. Nepper, and J. Levaux. (*Proc. First World Metallurgical Congress*, Sept., 1951, 241–246 : Amer. Soc. Met., 1952). The authors report that the Perrin process effects a 50–75% reduction in sulphur contents on basic open-hearth steels which normally run between 0·030 and 0·045% sulphur for a slag with a CaO/SiO_2 ratio of approx. 3. In the Perrin process, the steel is poured from the furnace on to specially prepared slag in the ladle. Operational results are given.—D. L. C. P.

Basis of Calculation of Charges for Basic Open-Hearth Furnaces for the Production of Ordinary Steels. P. Krynicki. (*Bol. Assoc. Brasil. Metais*, 1950, **6**, July, 241–253). [In Portuguese]. The author studies the necessity for calculating the charges in basic open-hearth furnaces. Production losses in the absence of these calculations are studied and optimum pig iron compositions for producing good and cheap steel are examined. The chemistry of slag formation and the part played by the slag and the oxygen of the ore are considered. Tables of data for hot metal and cold charges are given.—R. S.

Silicon Limits Oxygen Utilization in Iron for Carbon and Silicon Removal. W. R. Lysobey. (*J. Met.*, 1952, **4**, Aug., 819–820). The author briefly discusses the use of the oxygen lance in steelmaking and indicates that it is not practical for the removal of silicon, manganese, and carbon from blast-furnace iron. The degree of oxygen utilization appears to be influenced entirely by the concentration of silicon, the molten metal reacting with less oxygen as the silicon content increases.

Hot Rolling Behaviour of Resulphurized Steels Improved with Manganese Sulphide Additions. S. Feigenbaum. (*J. Met.*, 1952, **4**, Aug., 825). In the manufacture of high-sulphur free-cutting steel at the Jones and Laughlin Steel Corp., sulphur additions in the form of manganese sulphide in the ladle have been made in trial casts. Manganese and sulphur recoveries were about 80% and 85% respectively, and the hot-rolling behaviour of the steel was greatly improved.—G. F.

Desulphurization in Open Hearth Furnaces and Its Technical Interpretation. E. Altuna Zarraga. (*Inst. Hierro Acero*, 1952, **5**, May, 513–523). [In Spanish]. The author discusses, on the basis of his own experience at the Altos Hornos de Vizcaya, conditions which will facilitate desulphurization.

Scrap Handling Program Speeds Furnace Charging. W. F. Bowers. (*J. Met.*, 1952, **4**, Aug., 826–828). The author describes how the charging of furnaces at the Kaiser Steel Corp. has been speeded up by the development of two scrap yards, one dealing with the works' own scrap and the other with bought scrap. The various operations carried out in the two yards are outlined.—F. F.

Instrumentation Improves Quality of Finished Product. L. W. Heinzinger. (*Steel*, 1952, **130**, June 30, 90). **Steel Mill Instrumentation Improves Production and Quality.** L. W. Heinzinger. (*J. Met.*, 1952, **4**, Sept., 919–921). The author describes the instrumentation of the open-hearth furnaces, reheating furnaces, and rolling mills at the Seattle plant of the Bethlehem Pacific Coast Steel Corp. Increased production and yield, and improved quality are wholly or partly attributed to the installation of the instruments.—G. F.

Self-Baking Electrodes Spur Development of Bigger, Better Furnaces. M. O. Sem. (*Iron Age*, 1952, **170**, Aug. 28, 95–99). The development of electric furnaces subsequent to the introduction of the Söderberg electrode are described. Reliable operation of continuous self-baking electrodes has enabled the construction of larger, simpler and improved furnaces. Covered furnaces permit the collection of waste gases. Savings in coke and limestone are reported and lower electrode consumption and cleaner operations are possible.

Structural Repairs in Electric Arc Furnaces. V. Keclík. (*Hutnik* (Prague), 1952, **2**, 10, 214–220). [In Czech]. The problem of materials and methods for the repair of electric steel furnaces is considered, with special reference to roofs and linings.—P. F.

Effective Use of Electric Steelmaking Materials and Facilities. G. Soler. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 15–20). Methods by which electric furnace operators can conserve alloying elements under present conditions are discussed. The aspects considered include the utilization of alloy-containing scrap, grinding dust, scale, and alloy oxides; improved and uniform practices to promote higher yields and better alloy recoveries; and melting nearer the lower limits of the specification range.—G. F.

Boron in Steel—Principles and Practice. M. C. Udy. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 39–59). The author briefly traces the history of boron steels and then discusses in detail the effects of boron and correct practice in making such steels. The purpose of boron additions is solely to increase hardenability, and for optimum effect the steel should be fully deoxidized and fine-grained, additional aluminium, titanium, or zirconium being necessary. The optimum boron content is 0·001–0·003%, or somewhat higher in low-carbon heats (due to high oxygen) and electric furnace practice (due to high nitrogen). Within normal limits, boron additions to a given base material have no effect on tensile properties, temper brittleness, creep resistance at elevated temperatures, endurance properties, forging characteristics, welding characteristics, and corrosion resistance. The substitution of boron for other alloying elements is also considered.—G. F.

Boron in Electric-Furnace Melting—Special Considerations in Acid Practice. E. H. Berry, jun. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Steel Conf.*, 1951, **9**, 60–64). The author reviews the points, dealing with the manufacture of boron steels in acid electric furnaces, which were raised at the 1951 annual meeting of the Steel Founders' Society of America. The points dealt with particularly are the optimum boron content of steel, its effective addition, and avoiding boron loss during teeming.—G. F.

A Ladle Treatment for Desulphurizing and Degassing Steel. E. Spire. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 75–84). The author describes a method of desulphurizing and degassing steel in the ladle. The bottom of the ladle is fitted with a porous refractory plug through which argon is passed to agitate the metal in contact with the desulphurizing slag. Sulphur has been reduced from 0·035% to 0·020% on medium-carbon heats in industrial trials, and in addition, the treatment homogenizes the metal, imparts better temperature distribution, eliminates hydrogen and part of the nitrogen. The plug can be used for at least three heats.—G. F.

Chromite Hearth at South Works. J. E. Harrod. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 91–92). Operating experiences with a chromite hearth in the electric-furnace production of low-carbon stainless steel at the South Works of United States Steel Co. are described. Bottom life is found to be greater than that of normal basic bottoms.—G. F.

The Chromite Furnace Hearth. J. J. Green. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 93–94). The construction of a chromite electric-furnace hearth and operating experiences in the production of low-carbon stainless steels at the Bridgeville (Pa.) works of the Universal-Cyclops Steel Corporation, are described.—G. F.

Problems in Production of Stainless Steels Containing 0·03 Per Cent Carbon. R. B. Shaw. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 111–112). The author briefly reviews the electric furnace production of 0·03% C stainless steels at the Allegheny Ludlum Steel Corp., and outlines some of the problems peculiar to the process.—G. F.

Melting Low-Carbon Stainless Steel. J. H. Kennedy. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 133–114). A brief review is given of the practice employed at the Canton Works of the Republic Steel Corp. in the production of low carbon stainless steels.—G. F.

Basic Electric-Arc Furnace Metal for Casting Production. J. T. Gow and V. E. Belusko. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 174–192). The authors discuss the single-slag basic electric-furnace melting practices at the Electric Steel Foundry Co., and describe the means for control and the close control being obtained of the many variables in furnace practice and composition of the steels. The foundry produces castings of austenitic manganese steel, low-alloy heat-treatable steel, high-alloy stainless and nickel-base steels, and mild steels, and the fluidity obtained is at least equal to that obtained in acid practice. The conservation of scarce metals is also considered.—G. F.

The Basic Electric Melting Process for the Production of Cast Steel. A. Finlayson. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 193–209). The author describes in detail the basic electric melting process used at the Pacific Car and Foundry Co. for the production of steel castings. Particular attention is given to the basic lining, the charge, melting and the oxidation period, and the deoxidation and refining period.—G. F.

Acid *vs.* Basic Practice for Castings. I. W. Sharp. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 209–216). The Verona works of American Steel Foundries produces steel castings using both acid and basic electric furnaces. The author compares the two processes and shows that control of furnace operations is easier, and bottom troubles fewer, in basic practice. Comparison of operating costs is difficult because of differences in size and operating temperatures of the furnaces.—G. F.

Acid *vs.* Basic Steel for Castings. F. O. Lemmon and D. Helman. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 217–227). At the Springfield and Lima works of Ohio Steel Foundry Co., a comparison of acid and basic electric steel for castings has been made. It is shown that the basic practice effected no improvement in the casting quality, and owing to its higher cost the basic process thus cannot be justified at these works. It is felt that improvement in casting quality is more likely to result from careful study of mould condition and casting design.—G. F.

Basic Electric Furnace Practice at Crane Co. H. A. Young. (*Amer. Inst. Min. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 228–229). The author briefly describes the construction of the basic electric furnaces used at the Chicago works of Crane Co., and outlines the two-slag operating practice.—G. F.

Basic *vs.* Acid Practice for Steel Castings. S. F. Carter. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 230–234). A comparison has been made at the American Cast Iron Pipe Co. of acid and basic electric practice for the manufacture of steel castings, using furnaces identical in size and transformer capacity. The basic furnace is found to be more flexible and equally satisfactory for castings, but refractory and power costs are higher. For the manufacture of alloy steels, the basic practice can be justified economically because of higher alloy recovery.—G. F.

Electric Furnace Melting for Steel Castings. J. H. Hall. (*Foundry*, 1952, **80**, May, 160, 161, 318–336; June, 132–133, 276–283; July, 106–109, 239–246). The author traces the development of electric furnaces, and describes modern ones. He deals with lining the furnaces and charging, and explains the chemical reactions in basic operation. Acid practice and

the use of oxygen for faster heating are dealt with, and the coreless induction furnace is described.—E. T. L.

Effect of Manganese on the Activity of Sulphur in Liquid Iron and Iron-Carbon Alloys. J. P. Morris. (*Trans. Amer. Inst. Min. Met. Eng.*, **194** ; *J. Met.*, 1952, **4**, Sept., 939–940). The author describes the experimental procedure and results obtained in a study of the effect of manganese on the activity of sulphur in liquid iron and iron-carbon alloys. In the presence of manganese, the sulphur content of the metal increases at equilibrium and the activity coefficient of sulphur decreases, but at manganese contents below 3% the effect is small.—G. F.

Measurement of Viscosity of Synthetic Slags Rich in Ferrous Oxide. Parts I and II. G. Urbain. (*J. Chim. Phys.*, 1952, **49**, May, 308–315 ; 316–322). After presenting a short bibliography on methods of measuring viscosity which are applicable to metallurgical slags, an apparatus adapted for measurements on FeO-rich slags is described. The method of oscillation damping was chosen. The heating method and temperature measurement are described. Viscosity-temperature curves were determined for several synthetic slags related to the $FeO–SiO_2$ system. The information provided by the variations of viscosity and the cooling curves is in agreement with results obtained by other workers. In addition, isothermal curves of equal viscosity were obtained for those parts of the $FeO–SiO_2–TiO_2$ diagram that could be investigated with the available apparatus. (75 references).—T. E. D.

Continuous Casting of Steel Bars. (*Metalen*, 1952, **7**, July 15, 224–227). [In Dutch]. This is a brief survey of some of the difficulties occurring at the Beaver Falls, Pennsylvania, plant erected by the Republic Steel Corp. in conjunction with the Babcock and Wilcox Tube Co.—R. S.

Narrow-Side Vertical Cracking in Ten-Ton Ingot Moulds. A. Jackson and A. N. Whiting. (*J. Iron Steel Inst.*, 1953, **173**, Apr., 360–362). [This issue].

A Statistical Analysis on the Formation of Corner Cracks in Steel Ingots. A. Antonioli. (*Proc. First World Metallurgical Congress*, Sept., 1951, 298–320 ; Amer. Soc. Met., 1952). The author reports an investigation carried out when, in an electric steel melting shop pouring semi-hard steels, a high percentage of ingots developed corner cracks. A statistical analysis of melting and casting variables showed that corner cracks in steel ingots were due to too rapid a pouring rate and too high a pouring temperature. The analysis established the relation between these two factors.—D. L. C. P.

PRODUCTION OF FERRO-ALLOYS

Survey of the Extraction of the Most Important Steel-Alloying Metals. F. Johannsen. (*Metall u. Erz*, 1944, **41**, Apr., No. 7–8, 79–84). The necessary qualities of steelmaking additions are first described. Methods of manufacture of the principal alloying metals and ferroalloys are then outlined. Special reference is made to the use of low-grade ores.—L. H.

Production of Manganese from Iron Ores. G. Björling. (*Tekn. Tidskr.*, 1952, **82**, Sept. 9, 713–716). [In Swedish]. This is a review of methods for the economical production of manganese by smelting, including fractional reduction, blowing of spiegeleisen and matte smelting, and by leaching when employing SO_2, carbamate solutions, and electrolysis.—G. G. K.

Contribution to the Study of High-Percent Ferrosilicon. A. A. Trots and M. S. Maksimenko. (*Zhurnal Prikladnoi Khimii*, 1948, **21**, 7, 775–780). The following problems of the behaviour of ferrosilicon (75%) during processing were investigated : (1) Dust and gases formed during manual grinding and screening of ferrosilicon ; (2) gases evolved during boiling of ferrosilicon in water ; and (3) gases formed in a hermetically sealed mill during grinding of ferrosilicon. During grinding, 1 cu. m. of air contains 18·1 to 30·9 mg. and during screening 35·8 to 40·0 mg. of ferrosilicon dust. During manual grinding the air was found to contain 0·0003 mg./litre of hydrogen sulphide and 0·006 to 0·013 mg./litre of phosphine. Gas samples taken from the sealed mill showed 0·48–0·56 mg. of phosphine and 0·04 mg. of hydrogen sulphide per litre. The manual grinding and screening of ferrosilicon must be regarded as operations involving toxic hazards.—V. G.

FOUNDRY PRACTICE

Progress in Foundry Science in the Second Half of 1951. P. A. Heller. (*Stahl u. Eisen*, 1952, **72**, Aug. 28, 1094–1098 ; Sept. 11, 1169–1172). Progress in the metallurgy of cast iron is reviewed, dealing particularly with structures, properties, melting operations, fettling and mould making, the cast iron phase diagram, standardization, and testing.—J. P.

Progress and Development in the Production of Foundry Iron. J. Miles and F. Evans. (*Inst. Hierro Acero*, 1952, **5**, May, 471–475). [In Spanish]. The author briefly surveys developments in cupola blast control, mechanized charging, tapping and de-slagging methods, refractory linings and cooling, and, finally, duplex working with the cupola and the electric furnace.—R. S.

Design-Production Teamwork, Key to Foundry Engineering Advances. (*Amer. Foundryman*, 1952, **22**, Sept., 65–67 : *Canad. Metals*, 1952, July, 26–27). This report shows the benefits to be derived from co-operation between designers and foundrymen, known as 'engineering teamwork.' Examples of reduced costs of making tractor parts, trailer axles, and machinery hinge brackets are cited.—J. C. B.

Formation of Carbon Monoxide During the Passage of Carbon Dioxide over Glowing Coke at Various Temperatures and Speeds. G. Clas and E. Djernaes. (*Giesserei*, 1952, **39**, Sept. 18, 473–477). Conflicting views about combustion reactions in the cupola have led the authors to investigate the reducing processes during the passage of CO_2 through three different qualities of coke, electrically heated to 1200° C., the gas speed varying from 8·3 to 25 m./min. It is found that the rate of reduction to CO increases with temperature, decreases slightly with velocity (the temperature being constant) ; it does not depend on coke quality. In the cupola, reduction is confined to the first two or three layers of coke.—J. G. W.

Increasing Scrap Charging in the Cupola. L. H. Gross. (*Foundry*, 1952, **80**, Sept., 182–190). The author considers the achievement of the desired carbon equivalent in the casting, the method of fluxing, and the use of high proportions of scrap in the charge.—E. T. L.

Choice of Lining for Cupolas and Techniques of Ramming and Repairs. (*Fonderie*, 1952, Apr., 2903–2905). Factors influencing the choice between brick or rammed linings are considered. Methods of preparing and ramming monolithic linings are discussed, and methods of laying brick linings are described.—R. S.

Tapping into Several Ladles. W. Moll. (*Met. u. Giessereitechnik*, 1951, **1**, Dec., 358–360). A swivelling launder is described, the furnace end of which is supported on a large ball while the outer end is carried by a two-wheel bogie running on a track which is the arc of a circle. It is superior to any existing types of branched launder, which suffer from unequal division of flow between ladles, and oxidation of the metal during tapping on account of increased surface area.

Cupola Dust on Heating Surfaces of Heat Exchangers. A. Upmalis. (*B.W.K.*, 1952, **4**, May, 159–161). Experiments were carried out in two plants on the influence of dust on the performance of blast preheaters. The behaviour of the dust in the heat exchangers, its melting point and thermal conductivity were investigated. Means for the prevention of its deposition and for its removal from the heaters are examined.

Inoculation of Cast Irons. P. N. Bidalji. (*Hutnik* (Prague), 1952, **2**, 9, 187–189). [In Czech]. A lecture by Prof. Bidulja of the Metallurgical Faculty of Moscow University delivered to Czechoslovak metallurgists, dealing with the principles and mechanisms of the two basic types of inoculation for cast irons is reported. The thermodynamic aspects are considered, and Soviet developments in this field are reviewed, with particular reference to the author's own contributions.—P. F.

Cast Crankshafts and Camshafts. H. Jungbluth. (*Gjuteriet*, 1952, **42**, Aug., 123–129). [In Swedish]. A brief historical review of developments in crankshaft casting is followed by a review of suitable materials, including pearlitic and martensitic cast iron, Ford's alloy, and cast steel. Curves presented show that the fatigue, damping, and bearing qualities of cast iron are superior to those of cast steel. Illustrations of typical shafts are appended. Methods of camshaft casting are touched upon, the cams being hardened by chills, acetylene burners, and quenching, and by casting in green sand moulds. The author concludes that the unwillingness of designers to specify cast iron for such shafts indicates that casting is advisable only under certain conditions, with price as the governing factor, since, in many cases, forgings are so cheap that no advantage is obtained by casting.—G. G. K.

Mechanization of Shell Molding. (*Modern Metals*, 1952, **8,** Oct., 39–44). Developments in the Croning process in the U.S.A. over the last 2½ years are outlined. The use of sand and a thermosetting resin formed by means of a heated metal pattern plate permits a very accurate mold to be made. Castings with an accuracy of 0·003 in. are common, and often very little finishing is necessary. Savings in time, rejects and skilled labour are among the advantages.—T. E. D.

Casting Stainless Steel in Shell Molds. (*Foundry*, 1952, **80,** Sept., 108–111 : *Machinery*, 1952, **81,** Nov. 14, 1030–1032). The equipment and techniques by which Cooper Alloy Foundry Co., Hillside, New Jersey, produce cast stainless steel valves and pipe fittings by the shell mould process are described and illustrated.—E. T. L.

Gating for Shell Molding. W. A. Sokolosky. (*Foundry*, 1952, **80,** Aug., 92, 93). The author discusses briefly the use of strainers and chokes in gates for castings made by shell moulding.—E. T. L.

Sand Casting with Croning Process Shell Molds. R. Herold. (*Canad. Metals*, 1952, **15,** Apr. 21, 30–35). A review of the advantages of this process is given. The types of metal for which it is suitable are discussed. The difficulties encountered include coring problems, and use of resins and their cost. Shell moulding has enabled the foundry to recapture work from welding shops, and to compete effectively with forgings.—J.C.B.

The Croning Boxless Moulding Technique, a New German Moulding Method. F. Pölzguter. (*Giesserei*, 1952, **39,** Sept. 18, 467–472). Croning's method of dispensing with moulding boxes is described in detail. The preparation of the moulds is dealt with, and the economics of the process and the patent position are discussed.—J. G. W.

Investment Casting. R. W. Tindula. (*Foundry*, 1952, **80,** Sept. 193–208). The history of lost-wax casting, its definition, and uses are outlined. Descriptions are then given of moulding, patterns, and investment, and the precision possible and the limitations of the process are discussed. (73 references).—E. T. L.

Precision Casting and Preparing Moulds for It. H. Poetter. (*Met. u. Giessereitechn.*, 1951, **1,** Dec., 361–367). A comparison is made between five precision casting processes : (1) The frozen mercury process ; (2) the Antioch plaster process ; (3) the Bendix plaster process ; (4) the Croning plastic mould process ; and (5) the lost-wax process. The last of these is most suitable for materials which are not machinable after casting, particularly those with a high casting temperature. The frozen mercury process has been used for stainless steel, but is not recommended because of the large quantities of mercury required and danger to health.—H. R. M.

Metals Melted without Crucibles. E. K. Okress and D. N. Wroughton. (*Iron Age*, 1952, **170,** July 31, 83–86). A method for preparing high purity metals by levitation melting is described. In this method a metal is suspended by an electromagnetic field produced by coils surrounding but not touching the melt. When this is done in a vacuum, or in an inert gas, the melt is in suspension and is uncontaminated.—A. M. F.

Foundry Practice. J. F. Schnur. (*Metal Market Rev.*, 1952, **5,** Sept., 17–18). A number of suggestions for improving foundry practice are made ; these include higher quality patterns, improvements in simple hand tools, and the use of shell moulding.—B. G. B.

Blacking the Moulds and Cores in the Slag-Sand Casting Process. F. Naumann. (*Met. u. Giessereitechn.*, 1951, **1,** Dec., 372–373). Methods of applying blacking to intricate moulds and cores, particularly those having small holes, by means of a self-drying wash, are illustrated by examples.

Plaster and Plastics Combine in New Pattern Material. T. A. Dickinson. (*Foundry*, 1952, **80,** Sept., 174, 177, 180). Cheap plasters are cast, dried, and then coated or impregnated with polymerizable resin solutions. Furane resins are particularly suitable for this technique.—E. T. L.

The Growth of Cores in Difficult Grey Iron Castings. H. Kopp. (*Mitteilungen aus den Forschungsanstalten des Gutehoffnungshütte-Konzerns*, 1944, **10,** Apr., No. 6, 106–110). The causes of the undesired growth of cores in difficult castings were clarified by experiment, and the effect of core binder, sand, and the temperature was investigated. In carrying out experiments to impede core growth, the possibilities of reducing the longitudinal expansion of the core, using quartz sand and a higher casting temperature, were investigated. Works experiments substantiated the advantages of the proposed improvements.—R. J. W.

Pattern Making and Casting Design. F. Fischer. (*Giesserei*, 1952, **39,** Aug. 21, 411–413). This is a brief introductory article on the task of training operatives and the standardization of work in foundries.—J. G. W.

Pattern Production Methods. F. H. Wakeham. (*Foundry Trade J.*, 1952, **93,** Sept. 18, 331–332). The author deals with solid patterns, patterns joined with dowels, and patterns involving cores, taking a pump-valve box and a branch pipe as examples.—E. T. L.

Pattern Making Today. M. J. Kellner. (*Foundry*, 1952, **80,** Oct., 103, 293). The author traces the sequence of events in making patterns, considering the area of the casting, the material and its thickness, and actual fabrication.—E. T. L.

Special Mixture Overcomes Feeding Problem. (*Canad. Metals*, 1952, Sept., 30–32). This is a product of Foundry Services (Canada) Ltd. known as " Feedex 4 " which, when ignited, reacts chemically to produce heat. A ' trigger ' compound which initiates the heat-producing reaction and also a bonding material for producing solid moulds, are also used in conjunction with the mixture. Amongst the advantages claimed are reduction in feeder heads, elimination of rod feeding, and improved quality of castings.—J. C. B.

Swedish Nomenclature for Gating Systems. (*Gjuteriet*, 1952, **42,** Sept., 139–144). [In Swedish]. A draft list is published of suggested standard terms for gating systems. English equivalents are tabled, and some difficulties in classification are mentioned.—G. G. K.

Foundry Mechanization. A. S. Beech. (*Canad. Metals*, 1952, **15,** Mar. 34–39). An improved British sand slinger is compared with one used in the U.S.A., both from the economic and production aspects. The system of under-sand frame moulding, for the production of either deep or shallow moulds in large repetition quantities, is not properly understood in Canada or the U.S., and its advantages are stated. Finally, a release mechanism to avoid sand sticking in large hoppers is described.—J. C. B.

Supporting Cores in Grey Iron. A. Schmitz. (*Giesserei*, 1952, **39,** Oct. 2, 545–546). Practical problems arising during the location of cores are discussed and illustrated.—J. G. W.

Core Serves as Gate and Riser. B. F. Sweet. (*Foundry*, 1952, **80,** Sept., 96, 97, 236). A variation of the ' knock-off ' riser was developed for casting blank gears and sprocket wheels, up to 600 lb. in weight, in iron. Details of the technique are given.—E. T. L.

Purdue University Converts to Modernized Foundry. J. W. Mayer. (*Foundry*, 1952, **80,** Sept., 266, 268, 270). The new equipment in the modernized foundry at Purdue University is described and illustrated. The melting units for studying the casting of cast iron, steel, brass and aluminium, consist of a 2½-ton capacity cupola, direct and indirect arc furnaces, two induction furnaces, and a gas-fired furnace.—R. A. R.

Fork Truck Simplifies Handling in Quenching Steel Castings. (*Foundry*, 1952, **80,** Oct., 214, 216). The truck described lowers steel castings below floor level, and holds them during the quenching, thus playing a part in the treatment which is more than just transport.—E. T. L.

Your Quality Control Program—Is It Effective ? K. M. Smith. (*Foundry*, 1952, **80,** Oct., 188, 190–193). The author traces the history of the introduction of quality control into American foundries. Advice is offered on the status and function of a quality control supervisor.—E. T. L.

Circulation of Metal in the Mould Influences Quality of Castings. V. Oliverius. (*Hutnik* (Prague), 1952, **2,** 9, 190–193). [In Czech].

Modern Facilities Incorporated in New Alabama Pipe Foundry. W. G. Gude. (*Foundry*, 1952, **80,** Oct., 110, 113, 240). The new pipe foundry of the Alabama Pipe Co. is described. Hot metal is supplied from two balanced-blast cupolas, 66 in. inside dia., operated alternately. The desired composition of the iron is C 3·60, Si 1·60, S 0·10 max., P 0·90 max., and Mn 0·35%, and the pouring temp. is about 2400° F. The centrifugal casting machine mould is of alloy steel, and is coated with silica flour before casting each pipe. The annealing unit, 62 ft. is also described.—R. A. R.

Scabs—Their Causes and Cure. P. Nicholas. (*Fonderie*, 1952, June, 2983–2984 : *Foundry Trade J.*, 1952, **93,** Sept. 18, 326). A scab is due to the rapid expansion of silica transforming at 573° C., and it occurs chiefly when working with green sand. The cure is either to reduce the time of exposure to the heat, by increasing the size of the runners, or to use a

sand which allows some hot-deformation. Such a sand has a wide range of particle sizes.—E. T. L.

Preventing Shrinkage in Cut Gear Teeth. P. Dwyer. (*Foundry*, 1952, **80,** Aug., 205, 206). Practical recommendations are made for preventing shrinkage cavities and coarse structure at the base of teeth in machine cut cast-iron gears.

Simple Correlation in the Foundry. W. K. Bock. (*Foundry*, 1952, **80,** Aug., 104–109). This non-mathematical treatment illustrates the value of simple statistical techniques, taking as an example the correlation of impact resistance and hardness.

The Job Record Card, A Valuable Tool for the Jobbing Foundry. R. W. Griswold. (*Amer. Foundryman*, 1952, **22,** Sept., 34–36). The value of a job record card is discussed and some examples of suitable cards are given.—R. A. R.

Rational Production Management in Foundries. M. Förster. (*Giesserei*, 1952, **39,** Sept. 18, 498–504). Typical technical, personnel, and organizational problems of foundry management are discussed.—J. G. W.

The Engineer and The Foundry. D. C. Williams. (*Foundry*, 1952, **80,** Aug., 94, 259–266). The American engineer entering the foundry trade is described in terms of his professional outlook, need for reward, and for scope of his training.

The Central Laboratory of the Technical Centre for the Foundry Industry. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 161–167). The new laboratories recently inaugurated at Sèvres are described. The facilities cover all fields connected with foundry practice.—T. E. D.

Calculation of Casting Costs. (*Gjuteriet*, 1952, **42,** Aug., 130–133). [In Swedish]. A Swedish handbook published in 1945 dealing with costing for iron foundries has now been revised to take current price trends into account. New bases for establishing manufacturing costs are discussed. The application of a proposed system to a foundry (output 1000 tons/year) shows the calculation of metal and labour costs, overhead allocation and overhead recovery based on 80% and full production. Two alternative methods are given for the rapid pricing of castings.—G. G. K.

The Truth About Disease Caused by Foundry Dusts. O. A. Sander. (*Amer. Foundryman*, 1952, **22,** Sept., 53–54). In foundries, the airborne dust contains approx. 20% silica, causing a relatively slow silicosis. The remaining 80% modifies the disease, and makes the use of aluminium dusting unnecessary. Minimal silicosis is not disabling unless allied with tuberculosis or some conditions of old age.—E. T. L.

A Respiration Maintenance Program That Works. (*Foundry*, 1952, **80,** Oct., 194, 196). The system at an American foundry for issuing respirators and maintaining them is described.—E. T. L.

An Equipment Manufacturer Discusses Air Pollution Ordinances. J. A. Kane. (*Foundry*, 1952, **80,** Oct., 104–107, 204). The efficiencies of dust collection at various positions in a foundry, and American regulations on atmosphere pollution are discussed.—R. A. R.

HEATING FURNACES AND SOAKING PITS

Economies and Improvement of Quality in Forging and Heat-Treating. F. Barbas. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 37–41). The merits and disadvantages of the bell furnace are discussed. The horizontal furnace, the interior-muffle furnace, and the muffle furnace with an external gas seal are mentioned. The advantages of reheating without oxidation or in controlled atmospheres are outlined.

Recent Applications of High-Frequency Induction Heating in the Forging Industry. P. D. Jenkins. (*Metallurgia*, 1952, **46,** Sept., 119–124). The characteristics of the method of using H.F. induction equipment for heating materials before forging are outlined. Typical plant for heating billets, bar ends for upset forgings, and pins for bolt heading is described in detail.—B. G. B.

High-Geared Production with Induction Heating. (*Steel*, 1952, **131,** Sept. 8, 88–92). A new installation, at the Chevrolet works, for the induction heating of transmission gears before forging is described. Improved output, heat control, die life, working conditions, and cleanliness have resulted.—A. M. F.

Low Frequency Induction Heating. H. N. Acker. (*Canad. Metals*, 1952, Sept., 22–24). Equipment for induction heating of ferrous material, using 60-cycle current, has been introduced by the General Engineering Company of Toronto. This low-frequency induction heating may be used alone or in conjunction with higher frequencies. Its applications include forging of grinding balls and automotive parts, extrusion of non-ferrous alloys, tempering of components before drawing. This Company has also developed a self-excited generator for frequencies up to 1000 cycles which is easier to build than the conventional motor-generator set.—J. C. B.

Resistance Heating of Mild-Steel Containers at Power Frequencies. C. A. M. Thornton. (*Proc. Inst. Elect. Eng.*, 1952, Part II, **99,** Apr., 85–93). Methods are described of heating mild steel containers electrically, directly, at power frequencies, using the container itself as the resistance element. Two types of cases are discussed. A long container, such as a pipe, is usually heated by supplying a maximum of 25 V., conducted into the pipe. In the case of a short container, such as a vessel, it is more economical to employ inductive heating. Typical installations are described.—L. H.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Heat-Treatment of Alloy Steel Welds in Boiler Tubes. A. Krause. (*B.W.K.*, 1952, **4,** Feb., 44–47). The problem of normalizing welded seams in Cr-Mo heat-resisting steel tubes employed in steam generators, *in situ*, is considered. The application of several types of small portable gas-fired muffles is described and the use of photoelectric temperature-controlling equipment discussed. Metallurgical effects of the normalizing procedure are explained.—P. F.

Gas Carburising. (*Indust. Finishing*, 1952, **5,** Oct., 217–219 ; *Indust. Gas*, 1952, **16,** Oct., 8–12). The use of a Wild-Barfield gas preparation unit for removing organic sulphur from town's gas and increasing its CO content before using it for carburizing is described. The unit contains a special catalyst which can be regenerated. Equipment is available for gas flows of up to 200 cu. ft./hr., which will cope with charges of 15 cwt. in a retort of 2 ft. 3 in. in dia. by 4 ft. 6 in. deep. The advantages of gas carburizing over solid or liquid carburizing are considered.—T. E. D.

The Flame Hardening of Steel and Cast Iron. (*Usine Nouvelle*, 1952, Apr., 24, 25). This method of tempering by means of direct heat from a flame is essentially a superficial tempering, and the core of the material retains its orginal mechanical properties. Its advantages include its use in selective or partial tempering, lack of deformation during tempering, and low cost of equipment.—T. E. D.

Flame-Hardening of Entry-Edges of Steam Turbine Blades. F. J. Overkott. (*B.W.K.*, 1952, **4,** Feb., 47–49). Practical aspects of the flame-hardening of steam turbine blades, and questions of pre- and after-treatments are discussed. The methods are directly applicable to all steam turbine blades.

Increasing the Life of Rolls. O. Michailow. (*Met. u. Giessereitechn.*, 1952, **2,** Jan., 28–29). [Translated from *Trud.*, 1951, No. 114]. The process and equipment for the production of flame-hardened steel rolls in the U.S.S.R., as first developed at the Kusnetzker Kombinats Stalin and the Asowstahl Works, and the more recent installation at the Dsershinskijwerk Works, are described. Flame hardened rolls are giving 2 to 2½ times the life of unhardened steel or chill cast rolls.

Great Progress in the Maintenance of the Precision of Machine Tool Ways. E. B. Anpain. (*Acero y Energia*, 1952, **9,** May-June, 60–67). [In Spanish]. The results obtained with large flame-hardening machines used for hardening machine tool ways are reported and discussed.—R. A. R.

Reaction Equilibria During Protective Atmosphere Annealing. Part I.—Development of Protective Atmosphere Annealing and the Significance of the Gas Equilibria for the Protective Atmosphere's Properties. T. Berglund. (*Jernkontorets Ann.*, 1952, **136,** 8, 253–274). [In Swedish]. The author describes the development of gas annealing and the importance of the gas equilibria as affecting protective gas characteristics. Sections deal with : Carbon activity in steel as a function of temperature ; reactions in the Fe–C–CO–CO_2, Fe–H–O, and Fe–C–H_2–CH_4 systems ; influence of the water-gas reaction, equilibrium conditions between steel and gas ; influence of the Boudouardian reaction on gas composition ; and a brief review of Gurry's method of evaluating the characteristics of a protective gas. (25 references).—G. G. K.

Reaction Equilibria During Protective Atmosphere Annealing. Part II—Equilibria Calculations for the Gas Reactions in Protective Atmosphere Annealing. J. Berglund, S.-E. Erikson, B. Lindeskog, and R. Marstrander. (*Jernkontorets Ann.*, 1952, **136,** 8, 274–316). [In Swedish]. The authors deal with the calculation of gas equilibria in the Fe–C–O and Fe–H–O systems ; gas equilibrium with the

steel's phases in the Fe–C–H_2–CH_4 system ; the methane equilibrium, and the influence of the water-gas reaction on remaining reactions.—G. G. K.

Crack-Formation During the Heat-Treatment of a Tool. A. P. Gulyaev and S. P. Yakushev. (*Stanki i Instrument*, 1951, No. 8, 26–27). [In Russian]. Quenching experiments in oil and water on a number of tool steels are described. Specimens consisting of discs 10 mm. thick, 25 mm. in dia., with channels 8 mm. wide and 10 mm. deep cut radially, were used so as to reproduced the stress-concentrating shapes of many tools. The following conclusions were drawn : (1) Thermal stress alone does not cause cracking ; (2) cracking is very dependent on the temperature from which quenching takes place, and for each steel there is a critical temperature above which cracking frequently occurs ; and (3) the nature of the quenching medium and the rate of passing through the martensite range also influence cracking.—S. K.

FORGING, STAMPING, DRAWING, AND PRESSING

Pre-Determination of Pressure, Work and Power Requirements in Cold Upsetting. J. Billigmann. (*Stahl u. Eisen*, 1952, **72,** Sept. 25, 1221–1230). A simple table has been drawn up for the rapid calculation of pressure, work, and motor power requirements when cold upsetting of bolt heads, making use of certain approximation formulae. The pressure required is proportional to the square of the initial diameter of the working stock and increases exponentially with the amount of deformation. The work and power consumption, under constant upsetting conditions, increase as the cube of the stock diameter, and are also exponential functions of the amount of deformation, although the rise is not so rapid as with pressure. Conversion tables are provided to take into account the effect of different materials and shape of head. Nomograms are presented to enable the operating characteristics to be determined without calculation.—J. P.

The Technical Status of Forging Presses and Furnaces in Germany. A. Stodt. (*Stahl u. Eisen*, 1952, **72,** Sept. 11, 1131–1139). The design of modern German forging presses and ingot-heating furnaces are described and illustrated.—J. P.

Heavy Machine Foundations. R. Hammond. (*Machinery*, 1952, **81,** Nov. 21, 1094–1097, 1106). Vibration troubles are briefly discussed and an unusual foundation designed by J. H. A. Crockett for a forging hammer with an 8-ton tup owned by the International Nickel Co. Huntington, U.S.A., is described.—E. C. S.

Die-Contour Checking. (*Aircraft Prod.*, 1952, **14,** Dec., 432–434). A die-contour scriber is described and illustrated. This instrument is used to obtain the trace of, for example, a turbine blade die on glass record plates coated with a coloured film. A method of simultaneously projecting the traces of top and bottom dies is also described. It is a product of Optical Gaging Products, Inc., Rochester, N.Y., U.S.A., and the English agents are G. E. Marbaix, Ltd., London.—T. E. D.

New Stretch-Forming Techniques. F. Charity. (*Machinist*, 1952, **96,** Oct. 25, 1739–1741). With new stretch-forming techniques it is possible to form compound contours, produce larger unit parts, minimize trimming scrap, and to use soft, annealed, or half-hard sheet.—E. C. S.

Cartridge Cases are Drawn From Hot-Forged Cups. (*Machinist*, 1952, **96,** Oct. 25, 1745–1746). By deep drawing 75 mm. recoilless rifle cartridge cases from scale-free hot-forged cups instead of blanked discs, 14 progressive steps were eliminated, tons of spheroidized steel plate were saved, scrap handling was simplified, and tool and die cost was reduced.—E. C. S.

Hot Cupping, Cold Drawing Replace Machining Operations. G. F. Burnley. (*Iron Age*, 1952, **170,** Sept. 4, 152–153). A process for fabricating 81-mm. shells consists of hot forging billets, at 2000° F., in a 600-ton press. A backward extrusion is accomplished in three stages and the wall thickness of the shell is held to a tolerance of 0·015 in. Proper pickling and the use of a phosphate coating helps to obtain proper flow of metal during deformation. About 3 lb. of steel per shell is saved by this method of forming, and a costly machining operation is eliminated.—A. M. F.

Springback Problem in Metal Forming. F. Strasser. (*Steel*, 1952, **131,** Aug. 4, 90–91). The factors which determine the springback in metal forming or bending operations are briefly discussed. The methods for obtaining bends with exactly determined angles can be by overbending, restriking, or the use of special dies. Use of these dies should be avoided because the structure of the metal is unfavourably affected by their action.—A. M. F.

Hole Punching in the Sheetmetal Industry. R. Weisbeck. (*Canad. Metals*, 1952, **15,** May, 66–68). The factors affecting the production of clean holes in sheetmetal punching are reviewed. One of the difficulties is the stripping of the punch from the metal after the hole has been formed. To minimize this a spring stripper (the ' Strippit ') has been developed, and a self-contained hole-punching unit has been produced.

Tube Manipulation. (*Aircraft Prod.*, 1952, **14,** Oct., 326–331 ; Dec., 398–403). General methods of reducing, lapping, flanging, and doming tubes are described in detail, and the machines employed are illustrated. Draw-, compression-, roll-, press-, and ram-bending are similarly surveyed and illustrated.—T. E. D.

Cold-Drawing of Seamless Steel Tubes. B. Počta. (*Hutnik* (Prague), 1952, **2,** 9, 200–206). [In Czech). A detailed study is made of the principles and practice involved in the precision drawing of tubes from 0·14 to 5 in. in outside dia.—P. F.

The Ugine-Sejournet Extrusion Process. J. Strauss. (Amer. Iron Steel Inst.: *Blast Furn. Steel Plant*, 1952, **40,** Oct., 1184–1189). A review of the development and present status of this hot extrusion process which employs glass as a lubricant is made ; glass in fibrous or woven form is used. If the wear on the die is to be retarded, oxide must be removed from the surface of the billet ; this is at present done by carrying out the final heating in a salt bath. The production of most complex forms (some of them quite unbalanced) is now possible. Materials which are ' hot-short ' (*e.g.*, 29/9 Cr–Ni stainless steel), high purity molybdenum, heat-resistant alloys, titanium metal, and titanium alloys can be successfully extruded by this process.—B. G. B.

Deep Drawing Sheet Steel. E. M. H. Lips and F. J. H. Rolink. (*Proc. First World Metallurgical Congress*, Sept., 1951, 391–404 : Amer. Soc. Met., 1952). The drawability of metals can be predicted from tensile test data. It is shown that elongation and Erichsen values of the material are not decisive although the contraction in area is. The maximum attainable reduction is also determined by the yield value and the strain hardening parameter which may be derived from the tensile test. An improved die profile was developed, whereby the forming limit could be increased from 50 to 58%.

Calculations on the Influence of Friction and Die Geometry in Sheet Drawing. A. P. Green and R. Hill. (*J. Mech. Phys. Solids*, 1952, **1,** Oct., 31–36). Calculations of the drawing stress on the basis of the Hill-Tupper theory of sheet-drawing are reported for a range of reductions, die angles, and coefficients of friction. It is shown that, in frictionless drawing, the redundant work due to heterogeneity of deformation depends on a single geometric parameter.—J. G. W.

Phosphatization and Cold Working of Steel. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 105). The advantages of phosphatization in facilitating the cold working of steel are surveyed. A saving of as much as 20% in the cost of production of cold drawn tubes is attainable, and in wire drawing, much greater deformation in each pass is possible. Its advantage when stamping steel cartridge cases or the like is also described. Fewer operations and reheatings are necessary.

Modernization of an Old Draw-Bench for Cold Drawn Bars. N. L. Gripenberg. (*Proc. First World Metallurgical Congress.* Sept., 1951, 355–360 : Amer. Soc. Met., 1952).

Novel Press Setups Produce Cold-Drawn, Hot-Forged Parts. A. B. Mooers. (*Iron Age*, 1952, **170,** Aug. 14, 130–133). A method employed by Buick Motors to turn out rear-axle housing flanges consists of four cold-drawing operations in a single press, followed by two hot forging operations. In the cold drawing, slide feeds and automatic extractors are used for speed and safety.—A. M. F.

Squirting Cold Steel into End Product. (*Canad. Metals*, 1951, **14,** Dec., 50). The cold working of similar steel bars and billets to those used for conventional hot forging, into small industrial products such as shock absorber cylinders, gear blanks, and hub and axle components, has been achieved by Mullins Manufacturing Co. The " Koldflo " process is used which requires only standard hydraulic and mechanical presses. Large economies in scrap metal and costs are claimed.—J. C. B.

Comparison of the Technological Properties of Drawn Steel Wires as Affected by Patenting in Lead and Salt Baths and in Air. C. Eisenhuth and H. Krautmacher. (*Stahl u. Eisen*, 1952, **72,** Sept. 25, 1217–1221). Tensile, torsion and alter-

nating bend tests on carbon steel wires (0·5–0·8% C), reduced by 82% and 90% after patenting, have shown that lead and sodium nitrate baths are fully equivalent or superior to air patenting. The capital and running costs of sodium nitrate baths are considerably lower than those of lead baths.—J. P.

ROLLING-MILL PRACTICE

Simplified Determination of the Roll Force and Torque in the Rolling of Metals and Alloys. O. Emicke and K. H. Lucas. (*Met. u. Giessereitechn.*, 1952, **2,** Jan., 2–14). The existing formulae for the determination of roll force and torque in the rolling of steel are only of limited use to the practical roller and to the designers of rolling mills and mill motors, because they are complex when all factors are included and they are not directly applicable to non-ferrous metals. This paper gives nomograms for determining roll force and torque for cold-rolling of deep-drawing steel strips and a number of non-ferrous metals. They take into account roll diameter, strip dimensions, and the reduction of thickness in the pass considered and in previous passes.—H. R. M.

A Modern Rolling Mill Drive. (*Metallurgia*, 1952, **46,** Sept., 147–149 : *Eng. Boiler House Rev.*, 1952, **67,** Oct., 293–295). The recently installed rolling-mill plant at Richard Thomas and Baldwins Ltd., at Scunthorpe, is described. (*See J. Iron Steel Inst.*, 1953, **173,** Feb., 205).

The New Roughing Mill of the Donawitz Steelworks. O. Kammerhofer. (*Stahl u. Eisen*, 1952, **72,** Aug. 14, 1025–1034). The old roughing mill at Donawitz and the new plant and lay-out are fully described. The special problems associated with fitting American equipment to European conditions, and with the lack of space, are discussed.—J. P.

Wide Strip Mill Opens World of New Ideas. F. H. Fanning. (*Steel*, 1952, **131,** Sept. 8, 100–104). The development of the 4-high continuous hot and cold mills has rapidly displaced the hand operated sheet mills and has resulted in a very large increase in flat rolled products. Success in this process is due to prepared contour of rolls, composition and springiness of rolls, screw pressure applied to roll necks, and shape, composition and temperature of the strip.—A. M. F.

Control of Strip Thickness in Cold Rolling by Varying the Applied Tensions. R. B. Sims, J. A. Place, and P. R. A. Briggs. *J. Iron Steel Inst.*, 1953, **173,** Apr., 343–354). [This issue].

Works Trial of the 'T' Method of Automatic Gauge Control: A Progress Report. R. B. Sims, J. A. Place, and P. R. A. Briggs. (*J. Iron Steel Inst.*, 1953, **173,** Apr., 354–360). [This issue].

Roll Fractures in Mills for Rolling Thin Sheet. E. Gerlach. (*Met. u. Giessereitechn.*, 1952, **2,** Jan., 14–16). The causes of fracture of rolls used for hot-rolling are discussed. Recommended practices for their use and care, which are likely to minimize breakages, are given.—H. R. M.

Some Causes of Rejects in Plates. M. Borak. (*Met. u. Giessereitechn.*, 1952, **2,** Jan., 18–19). The causes of a number of different defects occurring in rolled plates, and the means of avoiding them, either in the rolling process or in the steel-making and teeming processes, are discussed.—H. R. M.

Strip Rolled to 0·0002 in. on New Small Mill. R. L. Stockus. (*Iron Age*, 1952, **170,** Sept. 4, 156–159). A description is given of a 4½-in. Sendzimir mill capable of taking 0·010-in. thick stock and reducing it to 0·0002 in. The work rolls are 0·250 in dia. and are supported by inside and outside backing and drive rolls. These are further supported by castor rolls within a solid housing. Constant tension control is provided by a balanced air loading system which can be preset to give values ranging from 10 to 400 lb. at speeds up to 300 ft./min.

Straightening Rolled Products. E. Siebel. (*Stahl u. Eisen*, 1952, **72,** Oct. 9, 1298–1301). Straightening is usually carried out by bending in the opposite direction to the existing curvature of the rolled material. The forces required and the resilience and internal stresses can be readily calculated. It is possible to explain the shortening observed after straightening bars, on the basis of the forces existing during the straightening operation. In contrast to bend-straightening, stretching induces no internal stresses, whereas straightening by local heating gives rise to the same stresses as does bending.—J. P.

MACHINERY FOR IRON AND STEEL PLANT

Diesel Cuts Scrap Yard Power Costs in Half. B. Waisman. (*Iron Age*, 1952, **170,** Aug. 28, 104–105). The installation of a 320-kW. diesel generator for use in a scrap yard is described briefly. Large variations in load caused by magnets and other equipment are quickly taken up.—A. M. F.

Traffic Control: Port Talbot, Margam, and Abbey Works. A. J. Harby. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 98–101). The author describes the organization and control of the road and rail traffic serving the Abbey, Margam, and Port Talbot Works of the Steel Company of Wales Ltd.—G. F.

Locomotive Haulage in Iron and Steel Works: Selection and Distribution of Locomotives. Part I—Locomotive Haulage in Shotton Works of John Summers & Sons, Limited. G. K. Tatterson. (*Iron Coal Trades Rev.*, 1952, **165,** Oct. 3, 739–741). The current expansion of the Shotton works of John Summers & Sons Ltd. is outlined, and locomotive haulage of hot metal, steel ingots, iron ore, coal, and other materials, is considered. The type of locomotive to be used for each purpose is indicated, and Diesel and steam engines are briefly compared.—G. F.

Locomotive Haulage in Iron and Steel Works: Selection and Distribution of Locomotives. Part 2.—Locomotive Selection for Iron and Steel Works. R. L. Crook. (*Iron Coal Trades Rev.*, 1952, **165,** Oct. 3, 741–743). The author first discusses the factors determining the power and weight of locomotives and then discusses the factors affecting the selection of locomotives for various duties within a steelworks. The advantages and disadvantages of Diesel and steam locomotives are considered.—G. F.

Traffic in Iron and Steel Works: Layout of Internal Rail Transport. M. D. J. Brisby. (*Iron Coal Trades Rev.*, 1952, **165,** Sept. 19, 629–635). The author examines some of the complex problems of supplies and services which call for the closest co-operation between rail transport and the iron and steel industry. He deals separately with inward, internal, and outward rail traffic, and discusses wagon turn-round, storage, and sidings capacity to meet all eventualities. Representative figures are given for consumption rates, tonnages hauled, and handling costs, and suggestions are put forward for fixing storage policy and siding capacities. Traffic organization and administration are also considered.

The Layout and Operation of a "Hump" Shunting Yard for a Foundry. A. Rüppel. (*Met. u. Giessereitechn.*, 1951, **1,** Dec., 376–379). A "hump" yard simplifies wagon shunting and economizes in track, locomotives, labour and operating costs. The layout plan and method of calculating the necessary gradients are described, and the suitability of various wagon braking devices is discussed.—H. R. M.

Supply of Oxygen for Metallurgical Trials. E. Karwat. (*Stahl u. Eisen*, 1950, **70,** No. 8, 336). Brief particulars are given of methods of storing oxygen, including containers and evaporators for liquid oxygen, and road vehicles with thin-walled insulated tanks carring from 2 to 5 tons of liquid oxygen.—R. A. R.

LUBRICANTS AND LUBRICATION

Methods for Expressing the Viscosity-Temperature Relationship of Lubricating Oils. J. F. T. Blott and C. G. Verver. (*J. Inst. Petroleum*, 1952, **38,** Apr., 193–249). The authors review critically the various systems and indices which have been proposed in the past for expressing, by a single figure, the viscosity-temperature susceptibility of lubricating oils. They propose a relative index, which is a modification of the present Viscosity Index, and for which the name Viscosity Modulus is suggested. Viscosity Modulus $= 100\times (\log_{10}L - \log_{10}U)/(\log_{10}L - \log_{10}H)$, U, L, and H being the kinematic viscosities at 100° F. of the oil under test and two standard series respectively.—B. G. B.

How Friden Applies Solid Film Lubricants. F. Shear. (*Machinist*, 1952, **96,** Nov. 8, 1828–1829). A new installation at the Friden Calculating Machine Co., Inc. applies a finish to steel parts that not only protects against rust but also gives excellent permanent lubricating properties to the surface.—E. C. S.

Boundary Lubricants. M. Kühn. (*Stahl u. Eisen*, 1952, **72.** Sept. 25, 1212–1216). The use of lubricants which adhere to the surface of metals by chemical bonding is described, and their action in fluid and mixed lubrication is clarified on the basis of the formation of orientated monomolecular layers. The lubricating action of these materials under high pressures and temperatures in boundary friction is explained as being due to the production of deformable glassy layers from the thermal decomposition products and their reaction with the substrate. Thus, long-chain phosphonic, phosphinic and arsonic acids can combine with the metal surface and their

long-chain organic groupings and provide suitable conditions for mixed lubrication. Under high temperatures, the resulting phosphoric and arsenic acids can combine with the metal to give phosphates and arsenates. Theoretical ideas about the nteraction of lubricant and carrier, and the relationship between chemical structure of the lubricant and diluent, and between lubricating power and adhesion are discussed. The most important sphere of application of such surface-active lubricants is in cold deformation.—J. P.

Lubricant Carriers for the Drawing of Low and High Carbon Steel Wires. W. Lueg and K.-H. Treptow. (*Stahl u. Eisen*, 1952, **72,** Sept. 25, 1207–1212). The various lubricant carriers are classified and their essential characteristics are specified. The production of lime and the effect of composition on its properties as a carrier are fully discussed. The advantages of borax over lime, and of the newly developed alkaline phosphate and silicate materials, are indicated. The use of films of colloidal graphite as lubricant carrier is briefly discussed; results indicate that the use of graphite will be restricted to special cases such as the drawing of tungsten wires.—J. P.

Metals in Oils; Their Determination by Spectrographic Methods and the Errors Involved. A. J. Ham, J. Noar and J. G. Reynolds. (*Analyst*, 1952, **77,** Nov., 766–773). The application of emission spectrography to the routine determination of metals soluble in lubricating oils is described. It is possible for one operator to analyse 17 samples per day for several constituents, with an accuracy, except for lead, comparable to that of routine chemical methods.—T. E. D.

WELDING AND FLAME-CUTTING

The Arc Welding of Low Chromium Molybdenium Steel Pipe. J. Bland, L. J. Privoznik, and F. J. Winsor. (*Welding J.*, 1952, **31,** Sept., 783–790). The arc welding of 0·5% Cr–0·5% Mo steel pipe has been investigated, using conventional cellulose-titania and low-hydrogen alloy steel electrodes. The Charpy keyhole-notch transition temperatures of the low-hydrogen-electrode welds were found to be below that of the as-received steel pipe material, but post-heat-treatment at 1050° F. per 8 hr. raised the transition temperature. No change in transition temperature, from that of the as-welded deposit, was produced by a post-heat-treatment at 1200° F. for 16 hr.

Chromium Recovery During Submerged-Arc Welding. J. G. Kerr and D. A. Elmer. (*Welding J.*, 1952, **31,** Sept., 431s–438s). Stainless steels are difficult to weld by the submerged arc process. Chromium is lost to the flux, and manganese and silicon are picked up by the weld metal. The data show that the extent of these reactions depend upon flux composition, and welding conditions. It was found that fluxes low in manganese and silicon and high in basic compounds are best for submerged-arc welding of high-chromium alloys.—U. E.

Submerged Arc Welding. O. M. Fromm. (*Iron Steel Eng.*, 1952, **29,** Sept., 100–106). The author deals with the semi-automatic submerged-arc welding process when used in conjunction with jigs and positioners, and describes the practice developed at the Sparrows Point plant of Bethlehem Steel Co. for maintenance and reclamation welding.—M. D. J. B.

Submerged Arc Welding at Atlas Steels, Ltd. C. Ekwall. (*Iron Steel Eng.*, 1952, **29,** Oct., 55–60). The author describes the submerged arc welding process and shows that these welds have excellent strength, ductility, uniformity, fatigue strength, density, and shock and corrosion resistance. They are very low in oxygen and nitrogen.—M. D. J. B.

Automatic Submerged Arc Welding Speeds Critical Repairs. (*Canad. Metals*, 1952, **15,** May, 48–50). This process has been applied to welding a hard alloy surface on to crawler tractor rollers and idlers. It is primarily for rebuilding critical track parts and it is possible to get good service from reclaimed items.—J. C. B.

The Continuous Cooling Transformation of Weld Heat-Affected Zones. W. R. Apblett, L. K. Poole, and W. S. Pellini. (*Welding J.*, 1952, **31,** Sept., 421s–430s). A new method is described for studying the continuous cooling transformations that occur in the heat-affected zone of welds. A tapered bar of small diameter, through which a heavy controlled current is passed to cause rapid heating and melting at the smallest section, is used to simulate the heat-affected zone of a weld. Almost exact correspondence between the weld thermal cycles and those of the tapered specimen is obtained. A number of specimens were quenched after fusion and a time-temperature-transformation diagram was constructed. Three distinctly different types of transformation were noted : (1) From homogeneous austenite which comprises the portion of the heat-affected zone nearest the line of fusion ; (2) from heterogeneous austenitic regions (heat-affected zone heated to 1700–1900° F.) ; and (3) from a field of high carbon austenite plus untransformed ferrite.—U. E.

Porosity in Mild Steel Weld Metal. D. Warren and R. D. Stout. (*Welding J.*, 1952, **31,** Sept., 406s–420s). The shielded inert-gas metallic arc or sigma welding process was used to study the effect on the formation of voids in weld metal of such variables as the deoxidation level of electrode and base plate, weld metal solidification rates, contamination on the surface of the base plate, and arc welding atmospheres.

Stresses in Circular-Patch Weld-Test Specimens. A. V. Levy and H. E. Kennedy. (*Welding J.*, 1952, **31,** Sept., 402s–405s). An investigation of the residual stresses in specimens of various plate sizes and patch sizes is reported. Patches of 3 to 8 in. in dia. were cut from plates 9 to 18 in. square and Union-melt welded to the plate. The residual stresses induced in the plates by the welding were determined with strain gauges. It was found that tensile stresses, both radial and circumferential, extended from the centre of the plate outwards beyond the weld zone for patch-diameter to plate-width ratios of 0·2 to 0·65. Residual tensile stresses of the order of the yield stress of the material were present in specimens of patch diameter to plate width ratio of 0·2 to 0·3. With increasing patch diameter the residual tensile stresses were reduced. Patches 4–5 in. in dia. seemed to make the best specimens.—U. E.

Should Preheat be Substituted for High Temperature Stress Relief in the Codes ? E. P. DeGarmo. (*Welding J.*, 1952, **31,** Sept., 393s–396s). The results of seven independent investigations show that a 400° F. preheat is just as effective as a 1200° F. stress-relief heat-treatment in improving the performance of low-carbon steel welds, welded with AWS Type E6010 electrodes, for static and impact load conditions. Preliminary studies with small specimens indicated that preheating may not be quite as effective as stress relieving in the case of fatigue.—U. E.

Observations on Experience with Welded Ships. D. P. Brown. (*Welding J.*, 1952, **31,** Sept., 765–782). A background is given of experience acquired with welded construction in the shipyards of the United States before and during the war construction programme.—U. E.

Welds in Machine Constructions. W. M. Halliday. (*Canad. Metals*, 1952, July, 40–42). This article outlines some of the chief physical and mechanical properties desirable in a weld metal to meet the stress conditions occurring in machine frames, cranes, runways, gantries, heavy jigs and fixtures for machinery.—J. C. B.

Welding Doubles Life of Forging Dies. A. L. Phillips. (*Steel*, 1952, **131,** Sept. 8, 86–87). It has been found that considerable savings can be made by building up a forging die block with weld metal. Originally the dies used were scrapped after three sinkings and had an output of 40,000 parts ; now the face can be built up eight times before scrapping, and the output has been doubled.—A. M. F.

The Technology of Automatic Welding in the Soviet Union. J. Mannin. (*Eng. Digest*, **13,** 1952, Nov., 387–392). The author presents a survey of automatic arc welding under granulated flux as practised in the U.S.S.R.

Welding Tool Steels with the Auswelding Method. T. Noren. (*Proc. First World Metallurgical Congress*, Sept., 1951, 412–419 : Amer. Soc. Met., 1952). An improved welding method for repairing or refacing tool steels is described. The method consists essentially in first heating the base metal to its proper austenitizing temperature, then cooling to, and welding at, the temperature at which the austenite decomposes most slowly. The technique is amply justified by the results obtained.

Welding Austenitic Steels for High-Pressure Boiler Plants. E. Kauhausen. (*Proc. First World Metallurgical Congress*, Sept., 1951, 429–435 : Amer. Soc. Met., 1952). The difficulties in welding austenitic steels for high temperature and high pressure applications in a boiler plant are detailed. A successful method of welding was evolved after careful tests for microfissures, corrosion, and embrittlement.—D. L. C. P.

Tensile Fatigue Strength of Plain and Welded Steel Specimens with and without the Influence of Corrosion. C. Appaly and F. Bollenrath. (*B.W.K.*, 1952, **4,** July, 223–226). For purposes of arc-welding seams on 0·12 in. Cr and Cr-Mn steel sheets, which were not subsequently machined or heat-

treated, unalloyed welding rods were found adequate. Specimens in which the welded area was machined flat on both sides were found superior in creep, fatigue and tensile strength, and corrosion properties to those from which the surplus weld metal was not removed.—P. F.

Influence of Heat-Treatment and Nitrogen on the Properties of Arc Welds. F. Gentner. (*B.W.K.*, 1952, **4,** Aug., 275–279). A report is given on initial experiments relating to the arc-welding of basic converter steel used in boiler construction. Mechanical properties of annealed, tempered and normalized welds are given. Normalizing is not recommended, tempering below 600° C. produced the best results from the point of view of tensile properties, ageing, and corrosion resistance. The beneficial effect of this heat-treatment could not be ascribed solely to the hardening effects of the carbon and manganese, but was traced to the influence of nitrogen.

Quality Control of Welds. W. Felix. (*Tecn. e Industr.*, 1952, **30,** Apr., 294–302). [In Spanish]. Characteristic essentials of weldments and the methods of examining them employed at the works of Sulzer Bros. are discussed.—R. S.

A Survey of Modern Theory on Welding and Weldability. D. Séférian. (*Sheet Metal Ind.*, 1952, **29,** May, 441–446; June, 529–532, 540; July, 627–634, 642; Sept., 827–832, 840). Conclusion of a series of articles (*see J. Iron Steel Inst.*, 1951, **169,** Nov., 299). Continuing the consideration of electrode coating, the author deals with refining effects of the coating, refining by reducing agents, the slag-metal equilibrium, and the manganese equilibrium. Acid slags, the equilibrium of basic slags, and oxygen distribution are next discussed, and the effect of the electrode coating on the mechanical properties of the weld, with special reference to the influence of carbon and hydrogen, is described. Finally, the manufacture of welding electrodes and coating by dipping and extrusion are outlined. Illustrations of equipment are given. (46 references).—P. M. C.

Notes on the Use of Radioactive Isotopes for the Industrial Radiography of Welds. J. Strebelle. (*Rev. de la Soudure*, 1952, **8,** 3, 145–152). The author reviews progress made since 1945 in the use of radioactive isotopes for the radiographic study of welds. The advantages of X-ray tubes over radium and radon, and the practical applications of synthetic radioactive isotopes are compared. Isotopes and γ-ray analysis present definite advantages over X-rays.—M. D. J. B.

Welding Tests on Steel with a High Yield Point Due to Submicroscopic Inclusions. K. A. Krekeler. (*Stahl u. Eisen*, 1952, **72,** Oct. 9, 1278–1280). The structural steel St 52 had to be produced, during the war, as a Si-Mn steel without other alloying additions, and difficulty was experienced in obtaining the desired yield point. Tensile and bend tests on welded specimens of a new low-alloy structural steel (St 50) with a higher yield point, due to submicroscopic inclusions produced by additions of aluminium, have shown that it can be welded successfully by electric arc, shielded arc, and submerged arc methods. The results of tensile tests on **V** and fillet welds were equal to those on a good St 52 steel. It was also shown that the new steel could be used satisfactorily with the same welding rods as St 52. The steel is superior to St 52 in respect of weld hardening and susceptibility to brittle fracture at low temperatures.—J. P.

Electric Resistance Flash Welding of Rails. R. Kühnel. (*Stahl u. Eisen*, 1952, **72,** Oct. 23, 1348–1349). Care must be taken to ensure that the ends of rails to be joined by electric resistance flash welding differ as little as possible in chemical composition and cross section. Using precautions described in a German Federal Railways specification for the selection and treatment of rail ends, perfect welds can be produced. A bend test is used for acceptance testing.—J. P.

Fabrication of Aircraft Undercarriages. (*Welding Metal Fab.*, 1952, **20,** July, 242–245; Dec., 443–445). A detailed illustrated account is given of the equipment and techniques employed to make gas pressure welds in aircraft undercarriage parts of Cr–Mo and Ni–Cr–Mo low-alloy steels. The microstructures and testing are also described.—R. A. R.

Testing Low Hydrogen Electrodes. S. Agnew, L. H. Stirling, and W. P. Campbell. (*Canad. Metals*, 1951, **14,** Dec., 40–46). In this technical survey by the Canadian Department of Mines, the incidence of hydrogen in welds was tested using low-hydrogen electrodes treated in different ways. The American Welding Society glycerine test was used. Baking the electrodes at temperatures from 150 to 500° C. reduced the hydrogen in the weld. Further, exposure of the electrodes to the severest humidity conditions in Canada increased the moisture absorbed in the electrodes, but it could be reduced by baking for 3 hr. at 120° C. Moisture-proof packaging of low-hydrogen electrodes in 10 lb. lots is recommended.—J. C. B.

Guided Bend Qualification Test. R. G. Alison. (*Canad. Metals*, 1952, **15,** Feb., 40–42). This test for qualifying welding operators is shown to be unfair.—J. C. B.

Industrial Brazing, Furnace Brazing. E. V. Beatson and H. R. Brooker. (*Welding Metal Fab.*, 1952, **20,** Aug., 293–298; Sept., 335–340). Copper brazing of ferrous metals is discussed, and metallurgical properties of the joint are given. Common types of brazing materials other than copper are also dealt with.—U. E.

Synthetic Resin Adhesives—Some Typical Examples of their Applications for Metal-to-Metal and Metal-to-Non-Metal Joints. (*Machinery*, 1952, **81,** Nov. 7, 975–982). Examples of strong joints between many different materials such as aluminium, stainless steel, cast iron, brass, rubber, ceramics, and glass, using Redux, developed by Aero Research Ltd., Duxford, are briefly described.—E. C. S.

An Electronic Tracing Head for Oxygen Cutting. H. E. Newton. (*Metro. Vick. Gaz.*, 1952, **24,** Sept., 227–231). The equipment described consists of a motor driven tracing head which is mounted on one arm of a pantograph. This is made to follow, by electronic means, an outline of the shape required executed in Indian ink on white paper. The usual templates are thus replaced by a simple drawing. In practice, the final product differs from the outline by less than 0·025 in.—B. G. B.

The Arcair Torch. (*Canad. Metals*, 1952, **15,** Jan. 26–27). This tool is for the foundry cleaning room, replacing the chipper's chisel and oxygen gouging torch. Compared with the pneumatic chisel, the arcair torch will cut out a defect at a given depth in much less time and the groove made is narrower, and therefore requires less weld metal to fill it up. The torch is $10\frac{1}{4}$ in. long and has a carbon electrode with air ducts running parallel to it. The air pressure required is 80–90 lb./sq. in.—J. C. B.

Pipe Fabrication with Oxy-acetylene. E. P. Auler. (*Canad. Metals*, 1952, **15,** Jan., 40–44). Three main uses of oxy-acetylene processes are given. In cutting operations with mechanical aids, it is possible to cut pipes at different angles to suit subsequent welding operations. For gouging, its merits are compared with grinding and chipping.—J. C. B.

Influence of Oxygen Purity on Speed and Gas Consumption in Flame Cutting. R. Gunnert. (*Tekn. Tidskr.*, 1952, **82,** Oct. 21, 865–870). [In Swedish]. Test strips 50 mm. wide were cut by machine, to ensure constant speed, from a 500 × 500 mm. plate, using standard nozzles placed 3 mm. from the plate surface. Oxygen of 97·6% and 99·8% purity was used. Plate thickness varied from 5 to 50 mm., gas consumption was 28 l./min. The increase in moisture in the gas when cylinder pressure drops has no effect on cutting. An increase of 1% in impurities in the oxygen raises gas consumption by 33%. Cutting speed decreases by 3–4% for each 1% increase in impurities. Fixed impurities in the actual plate are without influence on flame speed. Gas impurities are argon and nitrogen. Oxygen 100% pure would increase cutting rates by up to 30%.

The Cutting of Stainless Steels. J. Adriaans. (*Lastechn.*, 1952, **18,** Aug., 100–102). [In Dutch]. The author specifies the conditions required in a metal for oxy-acetylene cutting and indicates how stainless qualities fall short of these requirements. Aids to the cutting of stainless steels are mentioned; these include: (1) The Arcos hollow carbon steel electrodes through which oxygen is blown; (2) the Air Reduction process in which a fine powder of sodium salts is injected directly into the oxygen stream; (3) the Linde process in which a mixture of 98% iron, 1 to 1·5% calcium phosphate and 1 to 1·5% sodium bicarbonate is injected in a separate gas stream; (4) the Belgian oxy-kinetic process (addition of quartz sand to the oxygen stream to remove the oxides mechanically); and (5) the National Supply Company's carbon electrode with compressed air.—R. S.

MACHINING AND MACHINABILITY

Machining High Temperature Alloys. P. G. DeHuff and D. C. Goldberg. (*Machine Design*, 1952, **24,** Sept., 270–278). In both rough and finish machining, variations in any one of the following factors will have an immediate effect on the production rate and the quality of the completed part: (1) Inherent machinability of the alloy; (2) design of the part;

(3) design and type of cutting tool ; (4) capacity and rigidity of machine tool ; (5) accessory tooling ; (6) feeds and speeds ; and (7) coolants. Each of these factors is considered separately.—E. C. S.

"Cumulative Sampling" Boosts Shop Output 22%. C. H. Latter. (*Machinist*, 1952, **96**, Nov. 22, 1895–1897). Departing from conventional practice, Solar Aircraft combines process control and acceptance sampling in a single system. Scrap and remachining are 33% less, and machine-shop output is up 22%.—E.C.S.

Machining Research Reveals How Depth of Cut Affects Surface Finish and Hardness. P. Spear, I. R. Robinson, and K. J. B. Wolfe. (*Machinist*, 1952, **96**, Nov. 22, 1887–1894). Investigations have been made of the effect of depth of cut, feed, and speed on the surface condition of mild steel, 60/40 brass, and duralumin, when cut by high-speed steel, carbide, and aluminium oxide ceramic tools. X-ray, hardness and surface-finish methods were used.—E. C. S.

Radioactive Tools Permit Fast Tool Wear Studies. M. E. Merchant, H. Ernst, and E. J. Krabacher. (*Iron Age*, 1952, **170**, Sept. 4, 154–155). A method for measuring tool wear in machinability studies and other tests consists of making the tool radioactive and taking a count of the activity of the accumulated chips. Using this method, tests can be completed in a very much shorter time with accuracy comparable to the microscopic measurement of tool wear.—A. M. F.

Simple Screw Machine Form Tool Is 90% Cheaper. C. M. Woog. (*Iron Age*, 1952, **170**, Aug. 28, 102–103). The replacing of three dovetail form tools by two standard carbide tools in the machining of valve bodies in an automatic machine has effected considerable savings.—A. M. F.

Piercing Curvilinear Holes by the Electric Spark Method. V. I. Kurchenko. (*Stanki i Instrument*, 1951, No. 6, 34). [In Russian]. A very brief account is given of spark-machining equipment and technique which enable holes following curved paths to be pierced.—S. K.

New Technology for the Electric-Spark Method of Toughening a Cutting Tool. G. P. Ivanov. (*Stanki i Instrument*, 1951, No. 5, 20–23). [In Russian]. In the technique described, the durability of a tool is increased by making it the cathode, and a suitable material the anode, in a spark system at a maximum working voltage of 50. A certain quantity of the anode material, governed by a modification of Faraday's law, is deposited on and in the tool. Of the 13 anode materials tried, greatest increases in durability were obtained with anodes of ferro-chromium (70% Cr), various hard-alloys, or white iron (2·86% C, 1·0% Si). The effect was very marked when tool and anode were thinly coated with a plastic material. Machining and metallographic tests were used to find optimum conditions for the spark process and the best initial condition of the tool material.—S. K.

Shadow Method for Measuring Surface Quality. V. V. Bernavskii. (*Stanki i Instrument*, 1951, No. 8, 17–19). [In Russian]. In the method described, the profiles of the irregularities of a surface in the direction of the observer are exactly reproduced in the plane of the surface observed. This is accomplished by placing on the surface a glass plate with diamond-cut parallel straight lines, the shadows of the lines with illumination at 45° reproducing the profile. The lines are 1·5–2 μ thick, and the line spacing is chosen so as to be of the same order as the height of the irregularities. Compact equipment, with provision for photographic recording, is described.—S. K.

Some Recent Overseas Developments in Surface Finish. C. A. Gladman. (*Australasian Eng.*, 1952, Aug. 7, 61–71). The author discusses wear with boundary lubrication, the modern theory of friction being described in detail. The effect of finish on the load-carrying capacity of journal and thrust bearings is discussed, together with the relation between finish and the performance of piston rings and cylinders, the life of cutting tools, corrosion and fatigue of metals, and some other engineering functions. The mechanism of chip formation and the mechanics of metal cutting are next described, and the effect of cutting speed, tool shape, feed, and lubricant upon the finish of the workpiece is discussed in relation to the commoner machining processes. (57 references).—P. M. C.

Review of Surface Finish Literature. J. W. Sawyer. (*Machine Design*, 1952, **24**, Sept., 147, 290–307 ; Oct., 328–348 ; Nov., 286–294). The author presents some 220 abstracts from the literature on surface finish from 1945 to 1951 inclusive, together with a subject index.—R. A. R.

Experience and Results in Machining Rolls on a Roll Lathe of New Design. H. Opitz, J. Kob and H. Hucks. (*Stahl u. Eisen*, 1952, **72**, Sept. 11, 1125–1129). A new roll lathe has been designed and built to allow sintered carbide tools to work at their maximum efficiency. A Swedish firm reports that with this new lathe the time for preparing rolls is reduced by 50–60% as compared with former machines.—J. P.

Modern Methods for Machining Railway Wheel Discs and Solid Wheels. A. Kuhlewind. (*Stahl u. Eisen*, 1952, **72**, Sept. 11, 1129–1131). While the conventional machining of wheel discs and solid wheels requires three operations and a working time of 1 hr., the same job can be performed on an automatic lathe in two operations and in a quarter of the time.—J. P.

The Machinability of Unalloyed Steel in Fine Turning. G. Wagner. (*Stahl u. Eisen*, 1952, **72**, Sept. 11, 1144–1148). It has been shown by "life/tool-path" tests that, at conventional low cutting speeds, free cutting steels have better machinability than case-hardening and heat-treating steels, whereas at high speeds, free cutting steels are worse than non-sulphurized steels with equal carbon contents. Under correct testing conditions, the machinability of unsegregated free-cutting steels and non-sulphurized carbon steels decreases with increasing carbon content. In steels with more than 0·55% carbon, sulphur has no action in improving machinability. Relative indices of machinability have been determined which are independent of the cutting tool.—J. P.

Deformation of the Turning and the Rim Zone during Turning. W. Leyensetter. (*Stahl u. Eisen*, 1952, **72**, Sept. 11, 1139–1144). A turning is shorter than the length of cut ; this contraction depends on cutting speed, tool rake and tool material. When it is possible to use high cutting speeds, the contraction with a negative rake is the same as with a positive rake. This indicates that above a certain speed (about 300 m./min.) the shape of the cutting tool is of only subordinate importance in the deformation process. Microsections of the turned articles show that, with heavy contractions, there occur thick slip lines asymptotic to the surface and heavily deformed zones in the material. Large turning contractions correspond to marked deformation in the structure of the rim zone ; with smaller contractions, the angle of the slip lines with respect to the surface is larger, but their thickness is less.—J. P.

Toward Lower Grinding Costs. K. R. Eland. (*Canad. Metals*, 1952, **15**, June, 42). Modified aluminous oxide, through changes in manufacturing technique, has acquired some of the properties of silicon carbide, among which is the capacity to dissolve or diffuse readily in low-tensile materials, resulting in fast stock removal of the latter. Lion Grinding Wheels Ltd. "XA" aluminous oxide material has been made into grinding wheels, and replaces some silicon carbide wheels on grey iron, brass and aluminium.—J. C. B.

Trepanning of Forgings. F. W. Lucht. (*Steel Processing*, 1952, **38**, Aug., 394, 395, 398, 399). A trepanning cutter employing a tungsten carbide tip is described and fully illustrated diagrammatically. The advantages of trepanning over boring are discussed, and the manner in which a conventional deep-hole boring machine or lathe may be converted for trepanning is outlined.—P. M. C.

Machinability of Gray Iron by Constant-Pressure Lathe Tests. E. A. Loria. (*Foundry*, 1952, **80**, Aug., 194–200). Constant-pressure lathe tests on cast irons at the Battelle Memorial Institute to establish machinability indices are reported. Extensive trials on the effects of silicon carbide additions to the cupola indicated an overall improvement in machinability of about 5% to grey cast iron. Such additions also improved the machinability of high-phosphorus irons.

The Influence of the Inoculation of Grey Iron on Its Machinability. J. Kaczmarek. (*Przeglad Odlewnictwa*, 1952, **2**, 7–8, 239–249). [In Polish]. The factors influencing the machinability of pearlitic grey iron, inoculated with 0·3% and 0·5% of ferrosilicon, were investigated. From experimental evidence it is concluded that inoculation of cast iron has, in general, little influence on its machinability.—V. G.

Determination of Internal Stresses with the Help of Strain Gauges During Drilling and Turning. H. Bühler. (*Métaux-Corrosion-Indust.*, 1952, **27**, July–Aug., 325–328). A short account of the technique of the use of strain gauges for the determination of internal stresses is given. (13 references).

CLEANING AND PICKLING

Alkali Cleaner "Life". H. K. Hunt. (*Metal Finishing*, 1952, **50**, Oct., 72–73). The limits of cleaner life are deter-

mined by deterioration attributable to chemical and physical changes and not by any sudden breakdown of the cleaner chemicals. Thus, caustic soda is consumed by saponification of fats and greases and by absorption of CO_2 from the air, and wetting agents are removed by adsorption on solid particles of dirt and at solution-grease interfaces. Too high a content of dirt, grease, or oil, may be the result of employing too much emulsifier.—J. P.

Barrel Finishing Precision Parts for Jet Propulsion Units. D. J. Kaharl. (*Machinery*, 1952, **81,** Nov. 21, 1067–1073). Barrel finishing is a fast and economical method of deburring, smoothing, or improving the surface finish of parts by treating them in a revolving barrel containing water, a lubricating compound, and abrasive chips. The procedures employed at the Pratt and Whitney Aircraft Division, East Hartford, Conn., are described.—E. C. S.

The Importance of Surface Preparation before Painting: Shot Blasting. R. Marpon. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 97–99). The advantages of the use of shot instead of sand are explained. These include absence of dust, recovery of shot, and smaller quantity necessary for a given size of plant. Methods of preparation of surfaces are outlined, and systems for protecting shot-blasted sheet are discussed.—T. E. D.

An Experimental Study of Electrolytic Polishing. I. Epelboin. (*J. Chim. Phys.*, 1952, **49,** Feb., c214–c218). It is explained that the optimum conditions for electrolytic polishing are obtained with the electrolyte which gives rise to the greatest resistance maximum. It is also found that the composition of the bath is determined chiefly by the degree of hydration of the ions in solution. (10 references).—T. E. D.

The Water Content in Solutions for Electrolytic Polishing. I. Epelboin. (*J. Chim. Phys.*, 1952, **49,** July–Aug., c137–c141). Conditions of polishing using ClO_4' and PO_4'''' ions were studied. Optimum conditions for the polishing of various metals and alloys are represented on a ternary diagram, with the sides of the diagram representing the solvent, the water and the anion. The use of this diagram is explained. The mechanism of electrolytic polishing is discussed.—T. E. D.

Degreasing. (*Galvano*, 1951, **20,** Dec., 11, 12 ; 1952, **21,** Jan., 11–13 ; Feb., 11, 12 ; Mar., 11, 12). The problems of degreasing before pickling and electroplating, are discussed. The methods recommended include brushing, particularly with suspensions of lime, treatment in emulsifying and saponifying solutions, and in organic solvents.—J. P.

The Mathematical Treatment of Modern Continuous Pickling Processes. W. Fackert. (*Stahl u. Eisen*, 1952, **72,** Sept. 25, 1196–1207). The method described by Pearson, Bullough and Canning (*J. Iron Steel Inst.*, 1950, **165,** June, 185–190) for calculating liquor flow-rates in continuous strip pickling lines, amounts of water and ferrous sulphate to be removed, and acid to be added in a cyclic pickling-acid recovery process is reproduced. It is shown that, on a cost basis, the most favourable procedure for reclaiming the acid in spent pickle liquor, is one involving evaporation in order to precipitate ferrous sulphate as the monohydrate.—J. P.

Evaluating Surface Cleanliness. A. G. Gray. (*Products Finishing*, 1952, **17,** Oct., 58–80). Present researches aimed at evaluating the efficiency of cleaning procedures are summarized. Methods capable of detecting extremely small amounts of oil and grease include: (1) Measurement of contact angles of water drops ; (2) the water-break method ; (3) the water-spray pattern ; (4) radioactive tracers in the soil ; and, (5) the new American Electroplaters' Soc. atomizer test, which indicates the areas which are readily wettable and those which are not.—J. P.

PROTECTIVE COATINGS

Exposition of Principles Governing Electroplating. H. E. Arblaster. (*Australasian Plating Fin.*, 1951, **1,** June, 9–11 ; Aug., 14–17 ; Sept., 11–15 ; Dec., 3–9 ; 1952, Feb., 3–5 ; Mar., 8–13). The author explains in simple terms the scientific aspects of electrodeposition, dealing in Parts I and II with atomic structure and ionization, in Part III with rates of ionic migration and specific ionic mobilities, in Part IV with the electrolytic solution pressure of metals, in Part V with electrode potentials, in Part VI with polarization and the current required for plating baths, in Part VII with Glasstone's research on the electrolytic polarization of complex cyanides.

P.R. Plating—An Assessment of the Present Position. (*Electroplating*, 1952, **5,** July, 219–221). The advantages claimed for periodic reverse plating, particularly as applied to the deposition of copper, are the production of deposits which are somewhat smoother, of higher density and electrical conductivity, and of more even thickness over complicated shapes than those produced with straight direct current. The disadvantages are that the first cost of the plant is higher, the space requirements are greater, and the maintenance and electrical current costs are higher.—J. P.

Production Efficiency in Metal Finishing. H. Silman. (*Indust. Finishing*, 1952, **5,** Aug., 22–29 ; Sept., 136–139). The meaning and implications of industrial productivity under present-day conditions are discussed. Genuine increases in productivity in metal finishing may be achieved through improved shop layout, smooth flow production lines, efficient mechanical handling, improved vertical use of factory space, and the placing of small polishing or plating units, etc., in logical sequence in relation to other operations. Automatic polishing and plating plant should fulfil a complete cycle of operations without extraneous handling. New plant must allow flexibility of space and services for future expansions or modifications to processes.—E. E. J.

Apparatus for Adjusting the *p*H of Electroplating Baths. D. A. Vyakhirev. (*Zavodskaya Laboratoriya*, 1950, No. 5, 625–628). [In Russian]. Details are given of the construction, calibration, and use of a robust apparatus for measuring the electrolyt *p*H in plating baths. One electrode, of electrolytically-purified antimony, is continuously cleaned by a rotating brush, whilst the silver electrode is coated with silver chloride and immersed in saturated potassium chloride solution.—S. K.

The Coating of Metallic Surfaces by Electrolysis. P. Loyer. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 101). The qualities demanded of coatings are surveyed. The difficulties of surface preparation and of electrolytic methods are outlined, and the accepted standards of thickness are discussed. The thickness of layers of Cu–Ni before chromium plating is usually 40–50 μ in the U.S.A., while in France, 20 μ is considered sufficient for automobile parts. The minimum thickness of chromium which will take an acceptable polish is 0·3 μ.

Metallic Coatings for Surface Treatment. U. Trägårdh. (*Tekn. Tidskr.*, 1952, **82,** Sept. 9, 719–720). [In Swedish]. A review of corrosion protection methods by surface treatment of metals includes brief details of hot rolling of composite plates, hot galvanizing, metal spraying, and electroplating. Surface preparation is touched upon. A guide to the correct thickness of zinc coatings is given for various applications and it is stressed that contact with warm water over 60° C. renders zinc ineffectual.—G. G. K.

Testing the Quality of Electrolytic Coatings. P. Morisset. (*Rev. Gén. Méc.*, 1952, **36,** July, 195–201). Comments are made on French standards for testing electrolytic deposits. The nature of these coatings is described and the deposits are classified. Methods of testing adherence, continuity, and hardness, are discussed.—R. S.

A Contribution to the Study of the Poisoning of Hydrogen Electrodes. M. Bonnemay. (*J. Chim. Phys.*, 1952, **49,** July–Aug., c142–c146). An experimental study of the poisoning of a hydrogen electrode by electrolytic chlorine is described. The influence of overpotential on the degree of poisoning is used in explaining the part played by the electric field in the proximity of the electrode in the mechanism of the phenomenon.—T. E. D.

A New Polarographic Electrode. P. Lévêque. (*J. Chim. Phys.*, 1952, **49,** May, 269–272). This new electrode is composed of a platinum wire, pure or palladized, with an amalgamated surface, over which flows mercury. The surface, which is larger and less variable than that of the dropping mercury electrode, gives it some interesting analytical properties, which are described and discussed.—T. E. D.

On the Variations of Concentration at the Electrodes in Electrolysis: Inversion of the Hittorf Phenomenon. L.-H. Collet. (*J. Chim. Phys.*, 1952, **49,** July–Aug., c69–c71). Some experimental results obtained during the electrolysis of dilute hydrochloric acid between platinum electrodes showed that the ratio P_A/P_C of the anodic and cathodic loss of electrolyte, is positive at first, disappears between 10^{-1} and 10^{-2} N, becomes negative and tends towards a limiting value at large dilutions. This is a complete reversal of the Hittorf phenomenon.—T. E. D.

Graphical Representation of Aqueous Electrolysis Phenomena. P. Jolibois. (*J. Chim. Phys.*, 1952, **49,** July–Aug., c47–c48). A method of representing the state of progress of an electrolysis as a function of time is illustrated and explained.

Variation of Rate of Reaction of the Electrode with the Solvent. J. O'M. Bockris. (*J. Chim. Phys.*, 1952, **49,** July-Aug., c41–c46). Experimental results are surveyed, and parameters for the release of hydrogen in non-aqueous solutions are tabulated. Results are discussed. (42 references).

A Modern, Efficient Plating and Polishing Plant. F. M. Burt. (*Metal Finishing*, 1952, **50,** July, 57–63). The layout of the plant of the Crown City Plating Co. of Pasadena, California, and the type of work undertaken, are described and illustrated.—J. P.

Iron Plating of Electrotypes and Stereotypes. A. Lundbye. (*Electrotypers and Stereotypers J.*, 1952, **17,** July, 201–205, Aug., 233–234). This paper reviews iron plating of electrotypes and stereotypes at the research department, Kable Printing Co., Mt. Morris, Illinois. Details, including working temperatures and the current density most suitable, of iron baths containing chlorides, sulphates and mixtures of these salts are given. Important precautions to be taken are the correct bath *p*H, and scrupulous cleanliness, with no aeration to cause any oxidation of the salts to the ferric state. Under the correct conditions, copper and nickel plating of electrotype faces may be dispensed with.—J. C. B.

Cause and Removal of Faults in Jobbing Plating. H. W. Dettner. (*Metalloberfläche*, 1951, **3,** July, B97–B100).

Methods of Suspending Articles in Plating Tanks. O. Krämer. (*Metalloberfläche*, 1951, **3,** June, B81–B83). Descriptions are given of anode holders, and brackets and hangers for carrying articles in plating tanks.—R. A. R.

Beta-Particle and Gamma-Ray Thickness Gauges. M. G. Hammett and H. W. Finch. (*Electronic Eng.*, 1952, **24,** Nov., 536–539). The use of radioactive isotopes in measuring the thickness of sheet materials, or the coatings applied to them, is described. The material is passed between a suitable source and a detector, and as the transmitted β or γ radiation is proportional to the interposed mass, the thickness may be ascertained by measuring the detector signal. The most suitable detector is an ionization chamber with a suitable amplifier ; a description of a suitable circuit is given, with details of the installation and of the method of measuring coating thickness.—L. H.

Tests for the Control of Metallic Coatings. P. Morisset. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 87–91). The classification of coatings according to thickness is mentioned. Methods of measuring the thickness of electrolytic deposits and sprayed coatings are outlined, and tests for continuity and porosity of coatings are given. The importance of resistance to corrosion is emphasized and the efficiencies of various coatings are compared.—T. E. D.

Bright Chromizing by the French ONERA Process. B. Jousset. (*Proc. First World Metallurgical Congress*, Sept., 1951, 603–605 ; Amer. Soc. Met., 1952). In the ONERA process (Office National d'Etudes et de Recherches Aeronautiques) gaseous chromium fluoride decomposes at the surface of steel parts, the chromium diffuses into the metal, and the hydrofluoric acid recombines with chromium chips packed in with the steel. The mixture to form chromium fluoride consists of pulverized chromium, alumina, hydrofluoric acid, and ammonium chloride. It gives off CrF_2 at 100 to 250° C., and at about 900° C. its vapour pressure is high enough for the chromium to diffuse into the metal. In 1 hr. at 1100° C. chromium will give a layer 0·004 in. thick that is both homogeneous and ductile. On high carbon steel and cast iron, the chromized layer is less ductile but extremely hard because of the chromium carbide formed in the case.—R. A. R.

Some Special Features of the Chromium Plating of a Tool. D. G. Tufakov. (*Stanki i Instrument*, 1951, No. 5, 34–36). [In Russian]. The special features and optimum conditions for producing chromium deposits on cutting tools are considered. The following conditions are recommended : Concentration of chromic anhydride, 350 g./l., of sulphuric acid, 3·5 g./l. ; lead or lead-antimony anodes ; auxiliary thin-wire cathodes for some tool shapes ; temperature, 40–60° C. ; current density 20–55 amp./sq. dm. ; carefully controlled surface-quality of tool ; and thickness of deposit, 3–10 μ. The preparation, plating, and after-treatment are listed, and a table shows the appearance, causes, and means of preventing rejects.—S. K.

Hard Chromium Plating. R. de Buyer. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 93). Hard chromium plating, which is deposited at about 50° C., has a Brinell hardness of 800–1200. The particular advantages of this coating and its properties are surveyed.—T. E. D.

Anodes for Chromium Plating. E. R. Thews. (*Metalloberfläche*, 1951, **3,** June, B84–B86). The suitability of various metallic anodes for chromium plating is discussed. Lead containing 5–6% of antimony is shown to be the best.—P. F.

Material Control in Nickel Chrome Plating. R. T. Hood. (*Metal Finishing*, 1952, **50,** Sept., 64–66 ; *Metal Ind.*, 1952, **81,** Oct. 10, 287–288). A procedure is outlined for ascertaining plating costs, to provide a control for budgeting and estimating.—J. P.

Bright Nickel and Chrome Plating. F. Spicer. (*Indust. Finishing*, 1952, **5,** Oct., 192–194). Details are given of mass-production plating techniques specially devised for small articles having deep recesses.—K. E. J.

Protection of Steel from Radiation. R. Kopec. (*Hutnické Listy*, 1952, **7,** 8, 411–414). [In Czech]. The use of chromium diffusion into steel and the production of aluminium alloys in the surface of steels for protection against radiation are discussed, and methods by which this is done in Czechoslovakia are described.—P. F.

Hot Dip Aluminizing Affords Protective Surface. J. B. Russell. (*Steel*, 1952, **131,** July 14, 100–102). A process for the continuous coating of steel strip or wire has been developed. This involves degreasing, pickling, washing, preheating, coating, and cooling. The aluminium is held in a submerged electrode salt furnace where it floats on the surface and is heated by radiation and conduction from the salt. Ductility is primarily controlled by the thickness of the interfacial layer which is a function of the composition of the bath, time and temperature.—A. M. F.

White Brass Plating at Hamilton Mfg. Corp. C. C. Weekly. (*Metal Finishing*, 1952, **50,** Sept., 67–69). The use of a proprietary white brass plating solution to produce deposits suitable for replacing nickel on indoor furniture is discussed. The results of 15 million amp-hr. of plating over eight months have indicated that the process is economic, not difficult to control, and that the deposit has good mechanical and corrosion-resistant properties and will take a bright chromium finishing coat.—J. P.

Engineering Applications for Chromium Plating. R. Wick. (*Iron Steel Eng.*, 1952, **29,** Sept., 109–111). For most engineering applications it is claimed that finished chromium thicknesses of 0·006 to 0·015 in. are adequate and for particular cases up to 0·100 in. may be used profitably. This contrasts with decorative chromium plating 0·00001–0·00002 in. thick. The physical properties of chromium plate are discussed, specifications are given and the production of chromium plated machine elements is described.—M. D. J. B.

Comparative Wear Characteristics of Some Electrodeposited Metals. R. F. Ledford and E. A. Dominik. (*Plating*, 1952, **39,** Apr., 360–365). The effects of various additions to copper and nickel plating baths, and of chromium, on the wear resistance of some electrotyped panels have been measured. Results are shown graphically.—J. P.

Conservation of Nickel in the Plating of Chemical Process Equipment. A. K. Graham. (*Plating*, 1952, **39,** Apr., 369, 370, 376). Some methods, such as complete pickling, use of an iron deposit, and burnishing, to reduce the porosity of electrodeposited nickel coatings on steel are briefly discussed, the argument being that a thick coating can be replaced by a thin one if the latter has its porosity correspondingly reduced.

Late Developments in Electrodeposition Techniques. W. Blum. (*Canad. Metals*, 1952, **15,** May, 52–54). A review is made of the work done at the National Bureau of Standards on the physical properties of electrodeposited chromium and nickel. A method of improving the hardness of nickel by depositing it from a reducing solution is outlined in what is known as the 'Electroless' plating process. Tests for coat thickness, good adherence, and the stress in the deposit are given.—J. C. B.

Bright Nickel Deposits. R. Springer. (*Metalloberfläche*, 1951, **5,** July, A101–A110). This is a critical survey of methods of bright nickel plating; the relevant international patent literature is cited.—P. F.

Grain Size and Hardness of Nickel Deposits in Relation to Brightness. H. J. Read and R. Weil. (*Metalloberfläche*, 1951, **5,** July, A97–A101). It is shown that there is no relationship between brightness and the hardness and grain-size of nickel deposits.—P. F.

Nickel Anodes. E. R. Thews. (*Metalloberfläche*, 1951, **3,** July, B101–B107). The production and properties of rolled, cast, sintered and nickel-oxide-depolarized nickel

electrodes are discussed, with special reference to their performance in plating processes.—P. F.

A Modern Electro-Galvanizing Process. A. E. Williams. (*Indust. Finishing*, 1952, **5**, Sept., 116–121). Operating and technical details are given for the Bryanising process of continuous electro-galvanizing of steel wire, the British rights for which are jointly owned by British Ropes Ltd. and Rylands Bros. Sixteen lengths of wire are treated simultaneously and a perfect bond of the 99·999% zinc coating is achieved ; the coated wire may later be drawn to smaller sizes. Special points are the uniform coating thickness, control of deposition rate, and special preparation of the pure electrolyte.—K. E. J.

A New Electroplating Process "Promat." R. Zirilli. (*Pro-Metal*, 1952, **5**, Oct., 66–72). A new process for electroplating metallic objects with pure zinc is described. The process is very rapid—a coating of 12–15 μ is applied in 4–6 min. Details of the performance of the coatings are given ; the resistance to corrosion is excellent and the coating is very stable.—B. G. B.

Seldom Considered Sources of Trouble in Plating. G. Schuster. (*Metalloberfläche*, 1951, **3**, June, B87). Many troublesome effects, expensive to remedy, can be traced to faulty insulation of plating baths. Metal parts which have fallen to the bottom of the baths facilitate the passage of corroding currents through the bath walls in their vicinity.

Electrolytic Determination of Tin and Tin-Iron Alloy Coating Weights on Tin Plate. C. T. Kunze and A. R. Willey. (*J. Electrochem. Soc.*, 1952, **99**, Sept., 354–359). Tin and tin-iron alloy layers are stripped anodically from tinplate in HCl at constant current density. The changes in the potential of the specimen with respect to a reference electrode are recorded automatically as a function of time. The amount of current required to remove the various phases, as indicated by changes in potential, and thus their thicknesses, can be readily calculated by Faraday's Law.—J. P.

The American Society for Testing Materials Looks at the Metal Tin. B. W. Gonser. (*Metal Progress*, 1952, **62**, Sept., 72–75). A brief account of an A.S.T.M. symposium (New York, June 23, 1952) on tin is presented. A review of tin deposits and resources is first made. The use of tin for tinplate, in coatings of tin alloys, in the automobile industry, and methods of analysis of tin were discussed.—B. G. B.

Economical Finishing with Vacuum Metallizing. G. W. Carr. (*Metal Finishing*, 1952, **50**, Nov., 60–64). Bright and highly reflecting finishes may be applied to metal, plastics and glass surfaces by the newly developed vacuum metallizing process. Preliminary polishing of the articles is not obligatory because the necessary smoothness can be provided by a baked coat of alkyd lacquer. Large numbers of small articles can be loaded on jigs for insertion into the vacuum cylinder in which the coating metal, usually aluminium, is applied by evaporation. The articles require a further coat of lacquer to protect the thin metal layer from damage.—J. P.

The Mollerizing Treatment for Iron and Steel Surfaces. (*Machinery*, 1952, **81**, Nov. 21, 1073–1074). Aluminium can be united permanently with the surfaces of steel or iron parts, regardless of intricacy of shape, by the Mollerizing process in which the parts to be treated are submerged in an electrically heated salt bath after they have been thoroughly cleaned and pickled. The bath has a 2 to 4-in. layer of molten aluminium on the top. The parts are held in the salt, which is at about 1500° F., for a short time, and, as they are withdrawn through the layer, they receive a coating of aluminium about 0·005 in. thick. The salt bath consists primarily of barium chloride, and a positive direct-current electrode is submerged in the molten aluminium.—R. A. R.

Protective Coatings for Metals of High Melting Points. R. Kieffer and E. Nachtigall. (*Métaux–Corrosion–Indust.*, 1952, **27**, July–Aug., 312–317). Although molydenum, tungsten, tantalum, and niobum have very good mechanical properties at high temperature, they are very readily oxidised. The authors discuss methods which have been proposed to overcome this limitation ; these include the formation of surface coatings of the metallic borides or silicides. (34 references).

Touring a Silver Plating Plant. A. A. Parsons. (*Australasian Plating Fin.*, 1952, **2**, Apr., 3–9). The various stages in the plating process, as used in a modern medium-sized shop, are reviewed. Among the stages described are degreasing, pickling, 'quicking,' striking, plating, brightening, and the recovery of silver from spent solutions.—B. G. B.

Varnishes for Cans for Preserved Foodstuffs. F. Appell. (*Peintures, Pigments, Vernis*, 1952, **28**, Oct., 698–702). These varnishes must be corrosion preventatives, tasteless, odourless and non-poisonous, and their solvent must be rapidly and completely eliminated during baking. (49 references).

Information on Furnace-Dried Lacquers. H. E. Scheiber. (*Metalloberfläche*, 1951, **5**, Aug., A115–A117). Types, properties, and applications of furnace-dried lacquers are considered.

Sprayed Metal Coatings for the Protection of Steel Against Corrosion. J. C. Hudson. (*Electroplating*, 1952, **5**, July, 241–247). The performance of sprayed metal coatings on steel in exposure tests, the importance of thickness testing for specification purposes and the painting of sprayed metal coatings are discussed. The results employed are those obtained in the investigations of the Protective Coatings (Corrosion) Subcommittee of B.I.S.R.A.—J. P.

Prevention of Marine Corrosion by Metallizing Systems. H. Vanderpool. (*Welding J.*, 1952, **31**, Sept., 791–798). The metallizing method applied to steel for protection against marine corrosion is described. Metallizing requires the use of a machine which, with the aid of oxygen, acetylene and compressed air, converts the metal in the form of wire into small molten particles, and propels them at high velocity on to the steel surface. The advantages of applying metallizing systems to ships where a sprayed aluminium or zinc coating is covered with an organic top coat are discussed—U. E.

Some Modern Metal Spraying Pistols. H. Reininger. (*Metalloberfläche*, 1952, **6**, June ; *Electroplating*, 1952, **5**, Oct., 349–351). A number of modern metal-spraying pistols are described. These include the 'SMS' by H. Schlüpmann, Menden ; the 'Torpedo-Elektra' and the 'Torpedo-Universal' pistols by R. Fienemann ; the Colmonoy 'Spraywelder', made in Detroit ; the Metco pistols by Metallizing Engineering Co., Inc., New York ; and the 'Mogulectric' by the Metallizing Co. of America, Chicago.—R. A. R.

Metallizing and Its Functions. B. Simon. (*Iron Steel Eng.*, 1952, **29**, Sept., 112–113). The author describes metallizing as a rebuilding and production line process which saves material, time, and replacement costs in the fields of corrosion, rebuilding worn parts, and repairing errors in machining. Sand and grit blasting, rough threading, preparing hard surfaces and hard-surfacing techniques are described.

Fundamental Considerations in Stoving with Hot Air Ovens. (*Indust. Finishing*, 1952, **5**, Oct., 201–206). Means of improving efficiency of stoving ovens are discussed in terms of air temperature, humidity and velocity, flow and distribution of the air, and loading, positioning, and spacing of the work.

Rigid Production Control Assures Quality Enamelled Products. G. C. Close. (*Products Finishing*, 1952, **17**, Oct., 35–40). The fabrication, cleaning, pickling, and porcelain enamelling of reflectors and baths are described.—J. P.

A Method for Studying the Resistance of Enamels to Abrasion by Rapidly Moving Particles Suspended in High-Temperature Flames. N. E. Poulos. (*Bull. Amer. Ceram. Soc.*, 1952, **31**, Oct., 380–382). A laboratory procedure for testing the abrasion resistance of enamels to rapidly moving particles suspended in high-temperature flames is proposed.—E. C. S.

Islanding—A Surface Characteristic of Some Porcelain Enamels. M. K. Blanchard and W. A. Deringer. (*J. Amer. Ceram. Soc.*, 1952, **35**, Jan., 12–15).

Origin of "Fish Scales" on Enamelled Sheet Metal. J. Teindl and A. Havlík. (*Hutnické Listy*, 1952, **7**, 7, 344–347). [In Czech]. The conditions favouring the appearance of 'fish eyes' are examined and listed, and experimental work designed to test certain hypotheses as to their origin, carried out by the authors, is described. It is concluded that carbon monoxide and hydrogen in the furnace atmosphere during firing are detrimental, but that conclusive evidence as to the origin and nature of the formation of the above defect could be obtained only by X-ray stress analysis on samples with and without these 'fish eyes.'—P. F.

Applied Radioisotopes. (*Canad. Metals*, 1952, **15**, Feb., 36–39). A National Bureau of Standards investigation into the source of hydrogen which causes defects in porcelain-enamelled steel, indicated that water present in the frit used in the coating was to blame.—J. C. B.

Preliminary Study of the Compressive Stress Relief of Titania-Opacified Cover-Coat Enamels. D. C. Bowman and A. V. Sharon. (*J. Amer. Ceram. Soc.*, 1952, **35**, Nov., 300–303).

Ceramic Coatings for High Temperature Parts. (*Oil Eng. and Gas Turb.*, 1952, **20,** Mid-Oct., 225–227). A brief description of the Solaramic coating process developed by the Solar Aircraft Co., U.S.A. is given. The material to be coated is thoroughly cleaned and is often pickled. The coating is applied by spraying or dipping—one application usually being sufficient. After drying, the objects are fired at temperatures between 815° and 1090° C. for 10–30 min. Details of tests carried out on coated objects are given and they show excellent resistance to thermal shock, fatigue, and corrosion.—B. G. B.

Surface Treatments Preceding Enamelling. A. Pollack. (*Metalloberfläche*, 1951, **5,** Aug., A117–A118). Correct treatments and precautions necessary to prevent faulty enamelling are considered.—P. F.

An Electronic Finishing Process. J. Starr. (*Products Finishing*, 1952, **17,** Oct. 42–54). A process for spray painting and lacquering is discussed. The spray passes between electrodes maintained at high potential with respect to the conveyor-carried articles. The spray droplets collect ionized air molecules and are then attracted to the article to be coated.

Preventing Corrosion in the Plating Shop. (*Electroplating*, 1952, **5,** Oct., 335–336). Some general instructions are given for painting steelwork in plating shops.—J. P.

Painting and Lacquering in Industry. P. Wallgren. (*Tekn. Tidskr.*, 1952, **82,** Aug. 26, 681–684). [In Swedish]. The author discusses : Correct time for painting steelwork, surface preparation, choice of primer, climate's influence on paint thickness, measurement of paint film by electromagnetic meter, and the degree of rusting on untreated steelwork for periods up to one year. Swedish practice recommends a 40μ coating of primer on a smooth surface followed by a 40μ second coat. A 70μ primer should be applied to rough surfaces. In coastal areas the primer should contain 40% zinc chromate, in other cases 25%.—G. G. K.

Mica for Surface Coating. (*Indust. Finishing*, 1952, **5,** Sept., 130–135). The laminar structure of mica, and its large surface area, render it very suitable, when ground, for incorporation in corrosion-protection films, pigments and fillers. Tests show that it can improve the effects, and/or increase economy in use of anti-corrosive primer paints of several types ; reduce metal usage and settling rates in aluminium and bronze pigments ; and reduce settling rate and improve redispersibility in a wide range of conventional pigments.

Pretreatment Primers for Metal Work. E. E. Halls. (*Indust. Chem.*, 1952, **28,** Oct., 441–445). The author compares the results obtained with pretreatment primers with those given by other processes used to secure adhesion of paint and enamel finishes on metal surfaces.—R. A. R.

Paint-Spraying Apparatus and Equipment. H. Krause. (*Metalloberfläche*, 1951, **3,** Apr., B58–B60 ; June, B91–B92). Illustrated descriptions are given of immersion and spraying equipment for applying lacquer and paint to small articles and large surfaces.—R. A. R.

POWDER METALLURGY

Two Powder Metallurgical Methods to Prepare Alloy Specimens on a Laboratory Scale. J. Schramm. (*Proc. First World Metallurgical Congress*, Sept., 1951, 682–694 : Amer. Soc. Met., 1952). A method of preparing alloys by mixing powders of the constituent metals is described. The mixtures are pre-sintered and hot pressed. A second method involving impregnation of a sintered compact is also described. Many alloys have been made with a similar microstructure to the cast alloy.—D. L. C. P.

Grain Formation in Hard Metal Alloys. W. Dawihl. (*Microtechnic*, 1952, **6,** No. 3, 165–173). Practical experience has shown that with very fine grained alloy powders, of less than 3μ, the grain boundaries of cold pressed powder on heating pass through various intermediate stages. To investigate the nature of these changes, samples of cobalt, nickel, iron, titanium carbide and tungsten carbide were wet ground and the grain size examined by X-ray analysis. No noticeable increase in fineness was found except in the case of cobalt, which was reduced to less than 0·1μ, and transformed, probably, from the cubic to the hexagonal form.—L. H.

Effect of Particle Size on Certain Properties of Powdered Iron Parts. C. de Revoredo Barros. (*Bol. Assoc. Brasil. Metais*, 1952, **8,** Jan., 32–47). [In Portuguese]. The necessity for careful inspection of raw materials used in powder metallurgy is emphasized. The author studies the effect of particle size on some physical properties of sponge and electrolytic types of iron powder.—R. S.

A High Temperature Sintering Furnace. J. H. Rendall. (*J. Sci. Instruments*, 1952, **29,** Aug., 248–250). A laboratory furnace, intended for sintering refractory metal compacts, is described. The vertical electrical-resistance heater is a tube made of tungsten and molybdenum sheets, held in position by inner and outer spirals of tungsten wire, which also prevent distortion of the heater. The furnace is designed to minimize power consumption.—L. H.

Sintered Materials Resistant to Heat and Oxidation. R. Kieffer and F. Benesovsky. (*Métaux–Corrosion–Indust.*, 1952, **27,** July–Aug., 302–311). A review of the performance of some materials at high temperatures is made. It is considered that metallic alloys are now being used at their maximum temperature. Pure ceramic bodies are brittle and have low resistance to thermal shock. The development of composite materials, such as metal–ceramic bodies, or the borides, carbides, or silicides of metals is considered to be the most promising solution to the problems of materials resistant to heat and oxidation ; titanium carbide is very suitable material. The properties and performance of a number of sintered materials at high temperatures are given. (78 references).

Ingenious Tools Make Complex Metal-Powder Parts. G. H. De Groat. (*Machinist*, 1952, **96,** Oct. 25, 1731–1735). Some unusual pressing tools for making complex powdered-metal parts, and special sizing equipment for close tolerances, are described.—E. C. S.

The Possibilities of Improving Bearings of Sintered Metal Powders. T. Hövel. (*Stahl u. Eisen*, 1952, **72,** Sept. 11, 1157–1159). The properties of self-lubricating bearings of the usual sintered materials, particularly sintered iron, and the provision of oil channels are discussed. A possibility is the production of bearings by hot-pressing powders with large additions of graphite, which renders the use of oil lubrication unnecessary. Running-in troubles can be avoided with such bearings. Sintered iron has proved useful for supporting bushes because the bond between support and bearing metal is particularly good. A further development is the production of multi-face sleeve bearings, the principal advantage being the elimination of machining.—J. P.

Powder Metals. R. Talmage. (*Machine Design*, 1952, **24,** Oct., 145–149). Reasons for the rapid extension of the use of the powder metallurgy process are given, and materials, finishing, equipment and methods, design, and future trends are considered.—E. C. S.

PROPERTIES AND TESTS

New Factors in the Study of the Properties of Solids. F. Gatto. [*Met. Ital.*, 1952, **44,** Aug.–Sept., 391–398). The classical characteristics, such as the ultimate strength, reduction of area, and resilience, now tend to be replaced by new parameters of more general significance, which depend no longer on isolated quantities but on the general behaviour in the mechanical tests. A critical examination is made of these new parameters and their practical applications are discussed. (33 references).—M. D. J. B.

Instrumentation. R. H. Munch. (*Indust. Eng. Chem.*, 1952, **44,** July, 65A–68A). Bonded resistance wire strain gauges were originated to measure dimensional changes in mechanical structures ; their construction and operation, and means of increasing sensitivity and compensating for temperature effects, are described. They can be adapted to measure sensitively any variable capable of producing a small motion, *e.g.*, gas or liquid pressure, tension or compression loads, and torque.—K. E. J.

Test Bar Design Influences Results on Investment Castings. R. J. Wilcox. (*S.A.E. J.*, 1952, **60,** July, 32–36). Preliminary results are given of mechanical tests on both cast-to-size test bars and others machined from larger cast blocks. Four carbon steels, S.A.E. types 0022, 0035, 0050 and 0105 were examined. The design of test bar had a marked effect on the test results, the cast-to-size bars showing greater deviation in properties and usually lower values, particularly in reduction of area and elongation.—P. M. C.

A Practical Approach to the Theory of the 'Neutral Line'. L. H. Park. (*Sheet Metal Ind.*, 1952, **29,** July, 597–599, 606). The article deals with the effects of the metallurgical characteristics of the material, and the method of forming, on the position of the neutral line. Results are given of bend tests

carried out in order to check the relative merits of the formulae $\frac{\pi}{2}(0{\cdot}5t+R)$, $\frac{\pi}{2}(0{\cdot}4t+R)$, and $\frac{\pi}{2}0{\cdot}3t+R)$ for calculating bend allowances. (t = material thickness and R = bend radius). It is concluded that the formula $\frac{\pi}{2}(0{\cdot}4t+R)$ gives the best approximation for a 90° bend.—P. M. C.

Resilience Tests on Hard Steels. D. Giaccone. (*Met. Ital.*, 1952, **44,** Aug.–Sept., 399–421). The author describes tests carried out to see whether the criteria adopted in testing the resistance to shock of mild and semi-hard steels can also be extended to hardened hypereutectoid steels. Use was made of Mesnager test prieces without (or with very small) notches, and of unnotched Izod test pieces. Mesnager test without notch is definitely preferable, if more costly. It is shown that, in general, the passage from 64 to 63 and 62 Rockwell C hardness is followed by a gradual but appreciable increase in resilience, whilst for lower hardnesses this increase in resilience is not always apparent.—M. D. J. B.

Study of Impact Strength and Embrittling Factors. G. Remy. (*Métaux-Corrosion-Indust.*, 1952, **27,** June, 245–264). A detailed examination of the principal factors affecting impact strength and its study is presented. The result of variation in temperature on impact strength, and the effect of change in the velocity of impact and dimensions of the sample, are discussed. The effect on the impact-strength/temperature curves of variation in chemical composition or metallurgical state is shown. Work carried out on plastic deformation and rupture is reviewed. (46 references).—B. G. B.

Plastic Flow in a V-Notched Bar Pulled in Tension. E. H. Lee. (*J. Appl. Mech.*, 1952, **19,** Sept., 331–336). This theoretical analysis assumes a so-called Saint-Venant-Mises material, which flows at a constant yield limit given by the Mises criterion, and obeys the Mises flow-type relationship between stress and strain increments. The theory predicts the possibility of a plastic-flow type of rupture at a stress depending only upon the yield stress of the metal. The deformation normally encountered is deduced, and shown in a set of diagrams.—E. T. L.

Technical Principles in Plastic Deformation. G. Carro-Cao. (*Met. Ital.*, 1952, **44,** Aug.–Sept., 347–353). The author discusses the fundamental principles governing the plastic behaviour of homogeneous and isotropic metals in which plastic deformation takes place when the shear stress reaches a critical value. Based on the equations of equilibrium of the elastic theory and on experimental equations representing plastic behaviour, the author determines the various stress components and establishes the distribution of the velocities of plastic flow. (16 references).—M. D. J. B.

Fundamental Experiences on the Rupture of Solids. A. Guinier. (*Métaux-Corrosion-Indust.*, 1952, **27,** Apr., 150–155). A theoretical discussion is made on the rupture of ionic crystals and the cleavage of mica. Defects in solid structures, the Joffé effect, and Smekal cracks are discussed, and a brief account of some recent work on the rupture of metals is given.—B. G. B.

Theories of the Rupture of Metals. P. Laurent. (*Métaux-Corrosion-Indust.*, 1952, **27,** May, 189–200). A number of mechanisms proposed for the rupture of single crystals are first examined. The rupture of polycrystalline materials is next discussed, and the theory proposed by Ludwik is criticized owing to difficulties in defining and measuring experimentally the resistance to fracture. Work on rupture by fatigue and several theories to account for this are reviewed.—B. G. B.

Methods of Micro-Mechanical Testing. I. M. Roitman and Ya. B. Fridman. (*Zavodskaya Laboratoriya*, 1950, No. 5, 585–599). [In Russian]. This is a critical survey of the development and present state of the techniques of mechanical (mainly tensile) testing with specimens having maximum dimensions of the order of a few millimetres. Factors influencing the choice of the precise size and shape of specimen for a particular case are considered with special reference to grain size and structure, and machining conditions and cross-sectional area variations for specimens prepared on a miniature lathe are given. Successfully applied microtesting techniques are exemplified, as well as convenient methods of heating, quenching and polishing these small specimens. The special characteristics of microtesting machines are considered, and a machine designed by the authors is described. Investigation of the existence and influence of surface work-hardening on the microspecimens showed some effects for tempered structural steel, which could be removed by vacuum heating or electrolytic polishing, and no effects for hardened steel tempered at 500° C.—S. K.

Fracture Mechanism in Low Carbon Steels. Z. Kadeřávek. (*Hutnické Listy*, 1952, **7,** 7, 338–344). [In Czech]. The mechanism of fracture of some common low carbon steels was investigated by examining micrographs from the vicinity of the impact. Twinning in this region, and the nature of the cleavage of ferritic grains or groups of grains which were found characteristic of the first stage of fracture, were examined in detail. In fibrous fractures, this stage is accompanied by the formation of slip bands, without or with little, twinning. The reduction in cross-section arising from the crack induced by the impact leads to ordinary fracture by the spread of the crack. The rôle of pearlitic grains was not established, but they had no significant effect upon the fracturing process.

Brittle Fracture in Mild Steel. C. Schaub. (*Tekn. Tidskr.*, 1952, **82,** Sept. 9, 717–718). [In Swedish]. The tendency towards cracking in welded constructions is examined, with special reference to steel's sensitivity to welding which is affected by those alloying elements which increase hardenability and result in martensite formation. The carbon content should be limited to 0·20% and sulphur to 0·05%. Research during the last 10 years into cracking tendencies as governed by the characteristics of α-iron is briefly reviewed, including test methods by Tipper, Schnadt and Kuntze, and work by Robertson. Cracking may be due to sudden reorientation of some α crystals giving rise to twinning under high stresses, cohesion being lowered at the critical moment when all atoms in the affected area are in motion.—G. G. K.

Transverse Mechanical Properties in an SAE 1045 Forging Steel. A. H. Grobe, C. Wells, and R. F. Mehl. (*Amer. Soc. Met.*, 1952, *Preprint No.* 27). Tensile test data from specimens of a forging steel are analysed statistically. It is shown that the strength is increased by faster cooling from the annealed to the normalized state, although the ductility is unaffected. The transverse and longitudinal reduction of areas for samples from an ingot differed significantly, the latter being greater, but the direction of testing did not affect the tensile strength.—E. T. L.

Theory of Time-Dependent Rupture and Interpretation of Some Stress-Rupture Data. D. N. Frey. (*Amer. Soc. Test. Mat. Preprint No.* 80, 1952). Time-dependent rupture is shown to occur in two ways. At short times and high stresses the rupture is flow induced, and at long times and low stresses, where creep is negligible, rupture is of a truly brittle nature. The above theories explain the increased time of rupture, as a result of precipitation in austenitic alloys with rather large spacings ($>10^{-5}$cm.). Notch strengthening at short times follows from the restriction that such a notch has on creep and consequent dislocation condensation. Notch weakening at long times and low stresses follows from the stress concentration at the root of the notch. Materials having flow-induced fractures at both short and long times will be notch-strengthened at both time periods, whereas materials having true brittle fractures at long and short time will be uniformly notch-weakened.—B. G. B.

A Study of the Mechanism of the Delayed Yield Phenomenon. T. Vreeland, jun., D. S. Wood, and D. S. Clark. (*Amer. Soc. Met.*, 1952, *Preprint No.* 33). Plastic and anelastic microstrain of the order of 10^{-6} in./in. is observed before the onset of yielding in rapidly applied constant-stress tests and in repeated stress-pulse tests. Ageing of the specimens for a sufficiently long time allows recovery, thus preventing yielding. The activation energy of the recovery process is roughly that for the diffusion of carbon or nitrogen.—E. T. L.

Frequency Curves for the Yield Point of Structural Steel St. 37. G. Markiefka. (*Stahl u. Eisen*, 1952, **72,** Oct. 9, 1302–1303). Yield point results on 4232 specimens of an unalloyed hot-rolled steel St 37 have been plotted as frequency curves, as functions of the steelmaking process, and of the shape and thickness of specimen. These indicate to the structural engineer the minimum yield point likely to be encountered with this material.—J. P.

Load-Deflection Relationships in Slow-Bend Tests of Charpy V-Notch Specimens. R. Raring. (*Amer. Soc. Test. Mat. Preprint No.* 65, 1952). Load deflection diagrams of slow-bend tests at temperatures through the transition range of Charpy specimens of quenched-and-tempered and of pearlitic steels representing a carbon range of 0·01–0·63% and a

tensile strength range of 48,000 to 152,000 lb./sq. in. are shown. Results are compared on the basis of energy absorbed and of transition temperature, with those of impact tests of the same steels. Some correlation between the slow notch bend tests and tension tests is shown.—B. G. B.

How Good Is the Charpy Impact Test? C. A. Penrose. (*Iron Age*, 1952, **170,** Sept. 11, 143–147). Steels of 150,000 to 160,000 lb./sq. in. tensile strength have been used to determine the reliability and reproducibility of the **V**-notch Charpy test. A large number of test pieces having an impact strength of 20 to 30 ft.-lb. when tested at −40° F. gave a standard deviation of 1·8. From this it is suggested that it is unnecessary to report results to more than one decimal place. It was also found that the method of machining the test pieces had no influence on the results.—A. M. F.

An Inquiry into the Reproducibility of Impact Test Results. H. L. Fry. (*Amer. Soc. Test. Mat. Preprint No.* 67, 1952). The results of Charpy and Izod impact tests carried out using five different test machines are given. Large differences were found for the results obtained with the different machines on specimens of the same steel. Part of the error probably arises from insufficient clearance between the specimen supports and the striking edge, but if the machines were modified they would no longer conform to the present standards for the Charpy and Izod tests.—B. G. B.

Mutual Relationship between the Form of the Fracture of an Impact Specimen and the Mechanical Properties of the Steel. O. L. Bendryshev. (*Zavodskaya Laboratoriya*, 1950, No. 5, 609–614). [In Russian]. Comparative mechanical tests were carried out on specimens of two steels, one from an open-hearth, the other from an electric furnace, with the following respective compositions: C, 0·30, 0·33; Cr, 0·80, 0·82; S, 0·026, 0·006; P, 0·026, 0·016%. Composite specimens were oil-hardened from 890° C. and subjected to tempering at 220 or 510° C., followed by polishing. Both notched and smooth specimens were tested, macro-examination of the fractures revealing the frequent existence of stratification in the open-hearth steel, but seldom in the electrically-produced steel. Stratification tended to occur more frequently in specimens tempered at the lower temperature.—S. K.

A Study of the Grain Size of Steel. R. Calvo Rodes, P. Gomez Baeza and J. A. Garcia Poggio. (*Inst. Hierro Acero*, 1952, **5,** May, 530–574). [In Spanish]. The first part deals with a general theoretical study of the origin of the grain. Possible variations in grain size of heat-treated steels are examined. The effects of the potential grain size and the growth curve on the impact strength of a 0·4% carbon steel, and on the hardenability of a coarse and a fine-grained steel are considered.—R. S.

The Mechanics of Notch Brittle Practice. A. A. Wells. (*Welding Res.*, 1952, **6,** Aug., 68*r*–69*r*). The author presents his theory of notch brittle fracture. Strain hardening at the root of the notch, following the considerable increase of strain at the onset of general yield, is thought to be an important mechanism by which the stress is allowed to concentrate to a value sufficient to start brittle fracture.—U. E.

Experiments on Brittle Fracture of Steel Resulting from Residual Welding Stresses. R. Weck. (*Welding Res.*, 1952, **6,** Aug., 70*r*–82*r*). Preliminary results of an investigation on brittle fracture of steel following residual welding stresses are reported. It was found that when residual stresses were of sufficient severity, only spontaneous brittle failure may occur, and only when severe notches are present and the temperature is below the lowest value at which a severely notched tensile test piece would break in a ductile manner.—U. E.

Austenitic Cast Stainless Good for Low Temperature Applications—But. J. W. Juppenlatz. (*Iron Age*, 1952, **170,** Sept. 4, 147–151). The results of impact tests at temperatures down to −423° F. on a number of austentic Cr-Ni cast alloys are presented. These show that the ratio of austenite to ferrite or magnetic permeability are not related to impact strength at room or low temperature. Solution heat-treatment is preferred for best impact and corrosion resistance. Welding without subsequent heat-treatment can be performed. Impact tests at very low temperatures can indicate minor structural changes caused by previous heating in critical temperature ranges.—A. M. F.

On the Foot-Hills of the Plastic Range, H. W. Swift. (*J. Inst. Metals*, 1952, **20,** Nov., 103–120). An attempt is made to survey the activities of the several classes of pure and applied scientists engaged in the general field of metal plasticity. The present state of knowledge and the extent to which research is at present able to make its contribution to the various technological processes involving plastic deformation are discussed. A classification is suggested of the most profitable directions of inquiry appropriate to the mathematician, the metal physicist, and the engineering scientist, and an appeal is made for the co-ordination of programmes and for the presentation of reports and results in a form intelligible to other scientific workers. The methods of mathematical plasticity are illustrated by simple examples of shear line fields and flow grids. The limited success of more elementary methods of analysis is illustrated by the deep drawing of a cylindrical shell. An appeal is made for a more realistic model of lattice structure.—J. C. B.

The Application of Plastic Theory to Bending. M. Ish-Horowicz. (*Aircraft Eng.*, 1951, **23,** July, 203–206). General principles involved in applying "limit design" to structural members are discussed. Plastic theory, and the effects of the stress-strain function and shape of cross-sections on the moment of resistance are considered.—T. E. D.

Metals Can be Softened by Cold Work. N. H. Polakowski. (*Iron Age*, 1952, **170,** Aug. 28, 106–111). It is shown that cold working is a partially reversible effect and this is explained by applying the principles of the Bauschinger effect. Hard worked metals, even steels hardened by heat-treatment, may be softened by the application of cyclic deformation without being heated.—A. M. F.

Time-Temperature Dependence of Linear Viscoelastic Behaviour. F. Schwarzl and A. J. Staverman. (*J. Appl. Phys.*, 1952, **23,** Aug., 838–843). If a change of temperature is equivalent to a shift of the logarithmic time-scale, in connection with deformation, then the material is called 'thermo-rheologically simple'. The properties of such materials are discussed.—E. T. L.

X-ray Measurement of Residual Stress in Hardened High Carbon Steel. A. L. Christenson, and E. S. Rowland. (*Amer. Soc., Met.*, 1952, *Preprint No.* 22). An X-ray method is described in detail for the measurement of large stresses. Comparison of measurements of elastic strains on a hardened ring, by this method and by strain gauges, indicates a possible error of up to 4000 lb./sq. in.—E. T. L.

The Measurement of Internal Friction in Metallurgical Research. K. M. Entwistle. (*Proc. First World Metallurgical Congress*, Sept., 1951, 651–644 : Amer. Soc. Met., 1952).

The Counting Strain Gauge. J. B. Lambie. (*Engineering*, 1952, **174,** Aug. 8, 168–170). The author describes the uses and design of a counting extensometer developed by the Weybridge Division of Vickers-Armstrongs Ltd. The instrument, it is suggested, will contribute to the solution of the problem of assessing the life of aircraft structures. It will, by direct recording, contribute to the determinating of the fatigue properties of the materials and those of the structure.

Strain Rosette Analysis. H. Fealdman. (*Engineer*, 1952, **194,** Sept. 12, 341–342). The author describes methods of analysing the results obtained from multi-element wire resistance strain gauges in which the separate elements are variously orientated at a point. The theory underlying the Mohr's circle representation of two-dimensional stress and an analogous circle for strain is developed. Purely graphical methods for resolving the readings from two kinds of rosette, based on such circles, are then explained; the analytical solutions are also included.—M. D. J. B.

Stress-Probing: A Rapid Method for Stress-Surveying. N. Gross and P. H. R. Lane. (*Engineering*, 1952, **174,** July 25, 97–100). The authors describe a method of stress-probing pressure vessels by applying a commercially produced measuring gauge at predetermined points on the pressure vessel. The method described enables a comprehensive stress survey to be made about 20 times faster than with electrical resistance strain gauges. Comparisons between results obtained from strain-gauge measurements and stress-probing show very good agreement.—M. D. J. B.

Stress-Coat Examination of Strains in Sections of New Welded Plate Girders. R. Albrecht. (*Schweissen u. Schneiden*, 1952, **4,** Aug., 277–282). A study of strain distribution in welded plate girders by means of transparent plastic models and polarized light is reported.—U. E.

Thermal Stresses in Bodies Exhibiting Temperature-Dependent Elastic Properties. H. H. Hilton. (*J. Appl. Mech.*, 1952, **19,** Sept., 350–354). Expressions are derived for thermal stresses and strains due to a steady-state temperature gradient in a thick-walled cylinder, and in a thin circular plate, made of material having temperature-dependent elastic properties. Two examples are given which show that the maximum thermal stresses are smaller, although the maximum strains are larger, than for materials with temperature-independent properties. The solutions are only approximate.

The Effect of Quenching and Tempering on Residual Stresses in Manganese Oil-Hardening Tool Steel. H. J. Snyder. (*Amer. Soc. Met. Preprint* No. 21, 1952). The principal stresses in a fully hardened flat specimen are tensile, equal in all surface directions, and concentrated within a surface layer only 0·025 in. thick. These stresses can be reduced or eliminated by tempering. Etching in a 5% solution of nital introduces no stresses, but lapping induces compressive stresses to a depth of 0·0003 in., and grinding causes unequal biaxial stresses.—E. T. L.

Differentation Between Shrinkage and Fatigue Cracks in Steel Castings by X-Ray Analysis of the Oxide Layers Formed on the Fractured Surfaces. V. Montoro and E. Donati. (*Met. Ital.*, 1952, **44,** Aug.–Sept., 456). The authors examine a number of failures in steel castings and show how soon after failure the cracks exhibit zones of different colour. The darker zones are attributed to older shrinkage cracks and the bright zones to the fatigue failure. X-ray analysis of these two zones show the spectra to be different. The authors conclude that the failure fractures examined all started from shrinkage cracks.—M. D. J. B.

Some New Methods of Static Extensometry. H. de Leiris. (*Microtechnic*, 1952, **6,** 3, 155–164). Methods are outlined of tracing isostatic lines by covering the metallic surface to be tested with a brittle surface, and obtaining cracks in the surface by applying a load. The surface may be obtained either by applying a colophonium stick to the previously heated metal, or painting with a dissolved lacquer and allowing to dry. The correct type of coating is sufficiently brittle and possesses sufficient shrinkage stresses to produce orientated cracks with small additional extensions. The value of the results is discussed, and a theoretical treatment given. Finally, a suitable pneumatic extensometer is described.—L. H.

Law of Mechanical Similarity for the Deformation of Different Metals. G. P. Zaitsev. (*Zavodskaya Laboratoriya*, 1950, No. 5, 599–608). [In Russian]. In this mathematical study, similarity laws for different metals are worked out. After the general solution, plastic and elastic indentation by a sphere, and plastic and elastic extension are considered. Finally, the deformation of work-hardened metals is dealt with. It is concluded that, for similarity, two conditions must be satisfied : the dimensionless mechanical parameters of the two metals must be equal ; the quotients of the ratio of the load to the square of the linear dimensions divided by the dimensional mechanical constants must be equal. An equation is deduced for the indentation by a sphere of a specimen of finite surface.—S. K.

Fatigue—Its Nature and Some Ways of Reducing Its Incidence. P. L. L. Teed. (*Aircraft Prod.*, 1952, **14,** Nov., 362–365). A simplified survey of the subject of fatigue is presented. The engineering aspects of fatigue are discussed, and methods of reducing its incidence are described. (15 references).—T. E. D.

The British Welding Research Association. K. Winterton. (*Metallurgia*, 1952, **46,** Oct., 181–185). A fatigue testing laboratory has been opened at Abington providing facilities for testing full-scale structures and for other special investigations. Two new non-resonating fatigue testing machines of 100-tons capacity are described, together with other new equipment. Recent progress in engineering and metallurgical research is reviewed.—B. G. B.

Structure and Crystal State after Cold Deformation and Alternating Stressing. M. Hempel, H.-R. Sander, and H. M. Möller. (*Stahl u. Eisen*, 1952, **72,** Aug. 28, 1076–1087). If fatigue tests are supplemented by structural, chemical, physical and X-ray examinations, there is obtained not only the basis for the assessment of structural parts but also an insight into the processes occurring in the crystallites during the test. The structure and crystal state of an unalloyed 0·03% carbon steel were investigated after deformation under tensile load and alternating stress. The alternating stressing resulted in precipitations on the slip planes when the stress exceeded the fatigue limit and the number of cycles was greater than 2×10^5. The conditions under which precipitation occurs are discussed. Examination of residues showed that the precipitates consisted of cementite. Precipitation could be detected by X-ray methods as a sharpening of interference rings previously broadened by the tensile distortion.—J. P.

The Statistical Nature of the Fatigue Properties of SAE 4340 Steel Forgings. J. T. Ransom and R. F. Mehl. (*Amer. Soc. Test Mat. Preprint No.* 87, 1952). The statistical nature of the fatigue properties of SAE 4340 steel gun tubes was investigated using procedures which permitted statistical analysis of the variability of the endurance limits and life to failure. It was shown that the endurance limit and fracture curve of a heat-treated steel forging both exhibit variability. The variation in life to failure resulting from the variation in factors, such as the testing machine, eccentricity of rotation, stress relief, fracture criterion, stress calculation, and position of specimens in the forging, was shown to be insignificant.

A Statistical Interpretation of the Effect of Understressing on Fatigue Strength. E. Epremian and R. F. Mehl. (*Amer. Soc. Test. Mat. Preprint No.* 90, 1952). It was shown that for the two steels studied, the effect of understressing on fatigue can, in part if not entirely, be interpreted as a statistical phenomenon based on selectivity and the statistical nature of the endurance limit. Cold working does not play a dominant role in the behaviour.—B. G. B.

An Investigation of the Coaxing Effect in Fatigue of Metals. G. M. Sinclair. (*Amer. Soc. Test. Mat. Preprint No.* 92, 1952). The fatigue resistance of some metals may be improved by understressing followed by a process of gradually increasing the amplitude of the alternating stress in small increments, a procedure called "coaxing". In the present paper a study is made of the effect of various coaxing procedures on the fatigue resistance of ingot iron, SAE 1045 and 2340 steels, 75 S–T6 aluminium alloy, and annealed 70–30 brass. The results seem to indicate that the coaxing effect in fatigue is governed by a time-dependent localized strengthening through strain-ageing, and not by the ability of the metal to be strengthed by cold working.—B. G. B.

The Anisotropy of the Fatigue Properties of SAE 4340 Steel Forgings. J. T. Ransom and R. F. Mehl. (*Amer. Soc. Test. Mat. Preprint No.* 93, 1952). The anisotropy of fatigue properties in steel forgings was studied statistically using SAE 4340 steel from heats that showed both high and low transverse average reduction of area in tensile tests. The average transverse endurance limit was found to be 84% of the average longitudinal endurance limit in the high R.A.T. (reduction of area transverse) steel, while the ratio was only 68% in the low R.A.T. steel. Life to failure at the higher stresses was similarly anisotropic with a similar dependence on quality as judged by the reduction in transverse area. It was concluded that the anisotropy of fatigue properties is determined mainly by the presence of fragmented stringer-type inclusions.—B. G. B.

Corrosion Fatigue Resistance of Hollow Axles Increased by Pre-imposed Compressive Stresses. H. Buhler. (*Werkstatt u. Betrieb*, 1951, **84,** 474–466). The effect of residual compressive stresses in the surface zone of steel bars on their corrosion fatigue behaviour was investigated. If the bending fatigue limit of the furnace-cooled, and therefore stress-free, bar is regarded as 100, then prestressing (by quenching from 800° C.) increases the bending fatigue resistance under corrosive attack by 21%. The feasibility of the application of this treatment of large size machine parts was tested on the hollow axles of railway trucks with satisfactory results.—V. G.

The Endurance Limit of Temper-Brittle Steel. R. D. Chapman, and W. E. Jominy. (*Amer. Soc. Met.*, 1952, *Preprint No.* 23). Three different heat-treatments were applied to samples of SAE 5140 steel ; all produced the same hardness, but different degrees of brittleness. Fatigue, tensile, and Charpy impact tests at room temperature and at minus 35° F., showed that there was little difference between values of fatigue strength and notch sensitivity.—E. T. L.

The Influence of Different Surface Coatings on the Fatigue Strength of Steel. O. Forsman, and E. Lundin. (*Proc. First World Metallurgical Congress*, Sept., 1951, 606–612 : Amer. Soc. Mat., 1952). The results are given of fatigue tests carried out on a number of steels with and without surface coatings of zinc, tin, cadmium, nickel, and chromium, in air, tap water and 3% salt water.—D. L. C. P.

The Hardness of Metals. M. J. Druyvesteyn and M. A. Meyer. (*Metalen*, 1952, **7**, June 15, 203–206 ; July 15, 219–222). [In Dutch]. The interpretation of Brinell, Meyer, Vickers, and Rockwell hardness numbers is discussed and the hardness of single-phase alloys and of heterogeneous alloys (both annealed and chill cast) are dealt with. A new explanation is offered for the maximum hardness which occurs at the eutectic composition of a chill-cast alloy.—R. A. R.

Microhardness Testing. D. L. Paterson. (*Canad. Metals*, 1952, **15**, Jan., 46–51). The hardness procedure adopted by Canadian Westinghouse Co. using the Tukon microhardness tester is described. The applications include making a hardness traverse from weld metal to base metal.—J. C. B.

The Deformation of Austenite in Relation to the Hardness Characteristics of Steel. G. R. Bish and H. O'Neill. (*J. Iron Steel Inst.*, 1953, **173**, Apr., 398–405). [This issue].

What Price Hardenability? H. W. McQuaid. (*Metal Progress*, 1952, **62**, Sept., 88–96). An historical survey of the development and use of alloy steels (primarily in the American automobile industry) is made. Better design and finish of parts has resulted in the successful utilization of the lower priced and lower alloy grades of steel in conjunction with the highly developed heat-treating control now available.

The Applicability of Micro-hardness Testing as an Aid to Diagnosis in Ore Microscopy. J. Siebel. (*Metall u. Erz*, 1943, **40**, June, No. 11–12, 167–174). The use of microhardness testing of ore sections, as an aid to identification is discussed, particularly where similarities of structure, and other physical properties, might cause confusion. Conditions necessary are described, and sources of error indicated.—L. H.

A Hardness Tester for Small Loads. H. Broschke. (*Microtechnic*, 1952, **6**, No. 1, 15–21). The construction and method of use of the Leitz Durimet hardness tester is described. The range of loads is from 25 to 500 g., and it is free from lateral distortion. The magnification is 100 or 400 diameters. Provision is made for a camera attachment, and the instrument can also be used for scratch hardness tests.—L. H.

I.S.O. Agrees on Steel Tests. (*Steel*, 1952, **131**, July 7, 122–126). The International Organization for Standardization has agreed on test methods for Rockwell, Brinell and Vickers hardness tests, the bend test, and the Izod and Charpy impact tests. The tolerances on the various dimensions given in the tests have yet to be settled.—A. M. F.

Elimination of Yield Point Phenomena by Temper Rolling and Roller Levelling. N. H. Polakowski. (*Proc. First World Metallurgical Congress*, Sept., 1951, 553–571 : Amer. Soc. Met., 1952). The author describes his research on the above subject. A general explanation of the yield-point and ageing behaviours of temper-rolled, or roller-levelled steel sheet and strip is presented, based upon the work. The explanation also accounts for the effects of roll diameter and surface friction on the temper rolling process.—D. L. C. P.

Magnetic Measurement of Age-Hardening of Iron-Molybdenum Alloys. T. Mishima, R. R. Hasiguti, and Y. Kimura. (*Proc. First World Metallurgical Congress*, Sept., 1951, 656–667 : Amer. Soc. Met., 1952). The relation between mechanical hardness and coercive force of age hardening iron-molybdenum alloys was studied. The internal strain, as measured by means of magnetic properties, was correlated with the atomic process of age-hardening in these alloys.

Study on Age-Hardening. Y. Mishima. (*Proc. First World Metallurgical Congress*, Sept., 1951, 668–681 : Amer. Soc. Met., 1952). Data on age-hardening given in the literature are reviewed. The age-hardening of aluminium-copper, copper-silver and iron-copper alloys was studied.—D. L. C. P.

Rheotropic Embrittlement. E. J. Ripling. (*Amer. Soc. Test. Mat. Preprint No.* 73, 1952). The ductility deficiency exhibited by metals not crystallizing in the face-centred cubic system when these metals are strained at low temperatures, high strain rates, or in the presence of hydrostatic tension, has been shown recently to be partially strain curable. The portion of this embrittlement which can be overcome by prestrain under ductile conditions has been called ' rheotropic embrittlement '. The rheotropic behaviour of steel and zinc is considered and several variables which influence the behaviour are described and correlated.—B. G. B.

Effect of Hardness on the Level of the Impact Energy Curve for Temper Brittle and Unembrittled Steel. F. L. Carr, M. Goldman, L. D. Jaffe, and D. C. Buffum. (*Amer. Soc. Met.*, 1952, *Preprint No.* 26). Charpy impact tests on martensitic steel tempered to various hardnesses showed that, within the hardness range Rockwell C–27 to C–38 at least, there is a linear relationship between the hardness and the impact energy. The impact energy is about 10 ft.-lb. higher for unembrittled than for embrittled steel.—E. T. L.

The Effect of Various Heat-Treating Cycles Upon Temper-Brittleness. L. D. Jaffe, D. C. Buffum, and F. L. Carr. (*Amer. Soc. Met.*, 1952, *Preprint No.* 25). The increase in embrittlement of 20 specimens of SAE 3140 steel after various heat-treatments was determined in terms of the increase of transition temperature. When several embrittling treatments were applied to the same specimen, the nett embrittlement was often less than the sum of the separate effects. Quenching to room temperature between tempering and isothermal treatment had no great effect on the resulting embrittlement.

Temper-Brittleness of Steels. J. M. Vialle. (*Métaux–Corrosion–Indust.*, 1952, **27**, July–Aug., 281–301). A critical survey is presented of published work concerning temper brittleness of steels. Examples of the phenomena are given, and the effect of various factors (such as method of preparation, composition, structure, hardening, annealing) on the brittlnesse of steels is discussed. The following theories are reviewed : (1) The theory of reversible modification of special carbides ; (2) the theory of precipitation of one phase ; and (3) the hypothesis of intergranular segregation. (68 references).—B. G. B.

Effect of Arsenic and Antimony on Temper-Brittleness. G. W. Austin, A. R. Entwisle and G. C. Smith. (*J. Iron Steel Inst.*, 1953, **173**, Apr., 376–386). [This issue].

Temper-Brittleness in High-Purity Iron-Base Alloys. A. Preece and R. D. Carter. (*J. Iron Steel Inst.*, 1953, **173**, Apr., 387–398). [This issue].

Recent Modern Methods for the Prevention of Wear in Machines. F. Barbas. (*Usine Nouvelle*, 1952, May 8, 27–29 ; May 22, 27–30). The character of wear, and the effect of friction are outlined. The state of wearing surfaces is studied, and research on the mechanical activation of surfaces or the superficial layers of the metal by physical or physico-chemical methods is described. The importance of metalloids in suspension in bearing surfaces, including cementation with phosphorus or sulphur, is mentioned.—T. E. D.

Wear of Railway Materials—I. F. Šícha. (*Hutnické Listy*, 1952, **7**, 7, 356–364). [In Czech]. The present state of development of rails and of methods of testing their wear resistance are surveyed. Laboratory study of the microstructure and profile of rails which have been in service is reported.—P. F.

Thermal Cracking of Cast Iron Rolls. J. Thieme. (*Met. u. Giessereitechn.*, 1952, **2**, Jan., 17–18). Recent research has shown that some roll materials are more susceptible than others to initiation of heat cracks by thermal stresses. The development of these cracks, once formed, is largely due to mechanical stresses, and this also is influenced by the structure of the material. Although the occurrence of heat cracks depends on the working conditions, it can also be brought about by unsuitable treatment of rolls.—H. R. M.

Seam Fractures in Riveted Boiler Drums. S. D. Scorer. (*Elect. Times*, 1952, **122**, Nov. 6, 829–830). Although caustic embrittlement is considered to be the reason for a boiler explosion at York in 1949 it must not be overlooked that there are other causes of fracture. Any riveted seam, if sufficiently stressed, is likely to fail in one of several ways. In modern boilers there is little risk owing to improvements in design and construction, but in older boilers some parts may be slightly more stressed than others. The treatment of seam leakages and the testing of boilers by hydrostatic, radiographic, and magnetic crack detection methods are discussed.—B. G. B.

Abrasion and Wear. W. Späth. (*Metalloberfläche*, 1951, **5**, June, A81–A84). Methods of determining wear in processes involving slip and rolling, primarily of metal-to-metal systems, are discussed, and the question of appropriate parameters for the description of wear are considered.—P. F.

How Ferromagnetism Works. R. H. Esling. (*Iron Age*, 1952, **170**, Sept. 11, 148–152). The modern theory of ferromagnetism as explained by the theory of domains is described. The general principles of magnetism, flux density, hysteresis and eddy current losses are clearly explained.—A. M. F.

Magnetic Domains. H. J. Williams. (*Bell Lab. Record*, 1952, **30,** Oct., 385–396). A number of investigations concerning magnetic domains are discussed. The use of single crystals of iron enables the domain structure to be observed. The crystal is cut in the form of a hollow rectangle with sides parallel to the direction of easy magnetism and then a layer of a colloidal solution of magnetite in soap solution is placed over the crystal surface.—B. G. B.

Apparatus for the Measurement of Magnetic Susceptibilties. A. Pacault, A. Vankerckhoven, J. Hoarau, and J. Joussot-Dubien. (*J. Chim. Phys.*, 1952, **49,** Sept., 470–473). A translational pendulum of the Weiss-Foex-Forrer type is described and the design of the pole pieces is outlined. The study of the variation of magnetic susceptibility of certain diamagnetic and paramagnetic bodies with temperatures up to 1000° C. is possible. A force of 10–25 dynes can be measured to an accuracy of 0·02 dynes, corresponding to a displacement of 10^{-3} mm.—T. E. D.

A Wide-Range A.-C. Bridge Test for Magnetic Materials. D. C. Dieterly and C. E. Ward. (*Amer. Soc. Test. Mat. Bull.*, 1952, May, 75–80). The Hay bridge testing equipment and method are described. Experiments have shown that the Hay bridge circuit, when modified to compensate for copper loss in the test winding, is capable of giving accurate data on magnetic properties of Epstein samples or cores over a wide range of induction and frequency, with greater convenience of balancing and simplicity of calculation than other bridge circuits.—B. G. B.

Vibralloy—A New Ferromagnetic Alloy. M. E. Fine. (*Bell Lab. Record*, 1952, **30,** Sept., 345–348). This alloy was developed for use between −40° to +80° C. for reeds which are vibrated magnetically. It contains about 40% Ni with or without 9% Mo to give the correct temperature-modulus characteristics. The alloy is cold worked to obtain the desired mechanical properties and to leave sufficient permeability over the temperature range.—J. C. B.

New Ferromagnetic Materials of High Coercivity and Permeability and Small Losses. G. Guillaud. (*Rev. Tech. Luxembourg*, 1952, **44,** July–Sept., 137–154). A review of the magnetic properties of materials is first made. The manufacture of solid magnets by precipitation of two phases from a supersaturated solution (Comol, Cunife, Cunico, Vicalloy) or by diffusion (Alnico and Alcomax aeries) are discussed. Magnetic materials having a high permeability and based on iron–silicon, iron–nickel, iron–manganese and cobalt–nickel are reviewed. The general properties of magnetic materials based on a ferrite structure are described. A number of examples of materials having this structure are given ; they are used in transformers and magnetic amplifiers. Tables and graphs show the composition and magnetic properties of the large number of different alloys. (48 references).—B. G. B.

Magnetic Sorting. (*Aircraft Prod.*, 1952, **14,** Nov., 389–391). The application of a magnetic sorting, bridge-type instrument, manufactured by the General Electric Co., in which the hysteresis loop of a known specimen is compared with that of the unknown specimen, is described. Various wave forms obtained from specimens are illustrated and interpreted.—T. E. D.

A Process Permitting the Elimination of Parasitic Waves in the Ultrasonic Testing of Metal Parts. L. Beaujard. (*Compt. Rend.*, 1952, **235,** Oct. 13, 804–806). It is shown that parasitic echoes are due to surface waves which pass directly from emitter to receiver. Photographic integration permits their elimination without affecting echoes from actual faults.—A. G.

Ultrasonic Testing of Ingots: Techniques Employed at the Compagnie de Fives-Lille, France. M. Dehaine. (*Iron Coal Trades Rev.*, 1952, **165,** Oct. 3, 751–753). Details are given of ultrasonic methods of testing ingots which are used at the French steelworks Compagnie de Fives-Lille. It is shown that the method determines, with a satisfactory degree of accuracy, the position and extent of significant ingot defects, thus being of considerable value in obviating the rejection of large forgings at a late stage of processing.—G. F.

Ultrasonic Survey of Riveted Drums. W. W. Campbell and R. H. Murfitt. (*J. Inst. Fuel*, 1952, **25,** Sept., 190–195). An ultrasonic technique is described for detecting cracks between the rivet holes of the shell plates of riveted drums, where the presence of the butt straps precludes the use of other methods. The results of tests to check the method are given, and regarded as satisfactory for a developing technique.—D. L. C. P.

Ultrasonic Detector Reveals Flaws. W. J. Stirling. (*Canad. Metals*, 1952, **15,** Sept., 58–63). This detector is a radar set for locating flaws in metal. The author cites several uses for it in preference to X-rays for the non-destructive testing of metals. Amongst these uses are the testing of complicated aircraft components, railway engine axles, dies, and bond efficiency between babbit metal and its backing material on large thrust bearings.—J. C. B.

The Determination of the Elastic Constants of Metals by the Ultrasonic Pulse Technique. M. B. Reynolds. (*Amer. Soc. Met.*, 1952, *Preprint No.* 38). Values are given of the Young's modulus, shear modulus, and Poisson ratio for beryllium, niobium, zirconium, titanium, vanadium, thorium, uranium, alpha brass, and stainless steel. The technique and apparatus are described, and the computation of wave velocities is discussed.—E. T. L.

Putting Ultrasonics to Work. B. Wise and D. Ensminger. (*Product Eng.*, 1952, **23,** Aug., 180–185). Apparatus for producing sonic and ultrasonic vibrations and technical applications of these vibrations are reviewed. Because barium titanate can be moulded into many shapes and is electrostrictive in whatever direction an electric field is applied, it is very suitable as a transmitter of focused sound waves ; compared with quartz, a lower voltage is required to excite it. In addition to non-destructive testing of metals for defects, several other applications are very briefly outlined. These include (1) Dispersion of solids of low cohesion to produce semi-colloidal or colloidal particles ; (2) descaling and cleaning of metals ; and (3) drilling holes in hard materials and grinding in liquid suspension.—R. A. R.

Equipment for Non-Destructive Testing. G. Rigby. (*Canad. Metals*, 1952, **15,** May, 22–23). Three methods of testing are discussed. Magnetic particles are used in a Duovec which applies both currents in one operation ; and also in the Magnaglo which uses fluorescent particles. The Zyglo method, which uses a fluorescent penetrating oil, is described. The supersonic Reflectoscope for the detection of flaws deep inside a metal body is also described.—J. C. B.

A Torsion Pendulum of Improved Design for Measuring Damping Capacity. J. W. Jensen. (*Rev. Sci. Instruments*, 1952, **23,** Aug., 397–401). This instrument has been developed by the U.S. Bureau of Mines. Wide adaptability, accuracy, and simple and rapid operation are claimed for it. The design is calculated to minimize extraneous energy losses, *e.g.*, by friction, and harmonic and non-torsional vibration. The optical recording system is by light lever, recorded by a moving film camera with variable speeds. Examples of the use of the instrument are given.—L. H.

On the Determination of Adiabatic Elasticity. J. Perez Reyna. (*Inst. Hierro Acero*, 1952, **5,** May, 525–529). [In Spanish]. Using the magnetostriction properties of ferromagnetic materials, Pierce developed in 1928 a simple method of determining the frequency of a steel bar in which this frequency depends solely on the dimensions and velocity of sound propagation of the bar, and the propagation of the sound is a function of the modulus of elasticity and the density of the material.—R. S.

The Non-Destructive Investigation of Surface Defects. W. Stauffer and A. Keller. (*Escher-Wyss News*, 1950–51, **23–24,** 98–101). The magnetic powder, fluorescence and dye methods, for rendering surface defects visible, are outlined, and the use of a special lacquer as a developer in conjunction with these methods is described. The lacquer is transparent and colourless and can be removed as a thin skin after drying, thus providing a permanent document. The method is cheaper and more rapid than photography. Examples of records obtained are included.—T. E. D.

Surface-Crack Detection. (*Aircraft Prod.*, 1952, **14,** Dec., 431). A short outline is given of the Met-L-Chek method of surface crack detection, developed in the U.S. by Northrop Aircraft, Inc., and handled in Britain by C. J. Fox and Sons, Ltd., London. Cracks are revealed by coating the part with a penetrant red dye and developing with a white absorbent which shows up the defect as a large visible pattern.—T. E. D.

Some Simple Non-Destructive Methods of Testing Materials for Defects. J. van Nieuwkoop. (*Metalen*, 1952, **7,** Aug. 15, 247–254). [In Dutch].

Photoelectric Scanning of Fluorescent Indications. S. A. Wenk, K. D. Cooley and R. M. Kimmel. (*Non-Destructive Test.*, 1952, **11,** Summer, 28–31). Apparatus is described which was developed to detect automatically the fluorescent

indications received from flaws when using the magnetic particle or penetrant dye methods. A rotating mirror is used to reflect any yellow-green fluorescent emission into a photo-multiplier tube, which in turn passes a signal to a cathode-ray oscilloscope.—P. M. C.

The Metallurgist's Role in the Interpretation of Non-Destructive Testing. S. L. Henry. (*Non-Destructive Test.*, 1952, **11,** Summer, 16–20). The procedure used by the metallurgist to evaluate indications found by non-destructive test methods is outlined. Case histories are presented, covering: (*a*) Intergranular cracking caused by lead contamination shrinkage stresses on passing through the martensitic range; (*b*) non-metallic inclusions; (*c*) centre pipe; and (*d*) fatigue cracks in the columns of a 16,000 ton press.—P. M. C.

Non-Destructive Testing Personnel Problems. L. W. Ball and T. K. Chatham. (*Non-Destructive Test.*, 1952, **11,** Summer, 13–15). The authors discuss the desirable qualities and qualifications of personnel for non-destructive testing, and quote some of their experiences in operating a laboratory for research and development of this class of testing.—P. M. C.

Radioisotopes Aid Metallurgy. G. H. Guest. (*Canad. Metals*, 1952, **15,** Feb., 16–17, Mar., 18–19). The first part deals with a description of isotopes, their occurrence in nature, and the production of radioisotopes by bombardment. Their usefulness has been increased in recent years by more economic production and more sensitive electronic detection instruments to pick up low numbers of transformations. The second part indicates the uses to which radioactive isotopes may be put in metallurgical research. Chief amongst these are corrosion, diffusion, steel manufacture, flotation of minerals, and radiography.—J. C. B.

The Elements of Shipbuilding Radiography. K. V. Taylor. (*Trans. Inst. Marine Eng.*, 1952, **64,** Oct., 206–214). An account of the principles of radiography is first given, including details of the photographic methods, use of intensifying screens, and X-ray and gamma-ray sources. The detection of defects in welds and castings is discussed. The high penetrating power of gamma rays usually produces an image contrast markedly lower than that obtained with X-rays, so that images of flaws are more difficult to detect and the fine cracks may not be revealed. Gamma rays from iridium^{192} can be used with success on steel from $\frac{1}{2}$ to $2\frac{1}{2}$ in. thick.

New Sources of Radiation in Industrial Radiography. V. Vebersík. (*Hutnické Listy*, 1952, **7,** 7, 347–351). [In Czech]. The properties and manner of application of radioactive isotopes of cobalt, tantalum, and iridium for radiographic purposes are discussed.—P. F.

Using Iridium 192 for Castings Inspection. (*Canad. Metals*, 1952, **15,** Mar., 41). Iridium^{192} is being used to examine castings for flaws and faults and record them on the film. It has been investigated as supplementary apparatus to X-ray equipment for radiographic examination of heavy castings. The safety precautions are described.—J. C. B.

A Preliminary Report on the Picker-Polaroid Process in Industrial Radiography. J. A. Reynolds. (*Non-Destructive Test.*, 1952, **11,** Summer, 24–27). The invention by E. H. Land of the process which gives a finished photograph 1 min. after exposing a negative has now been successfully applied to radiography. A composite pack containing a negative, a pod filled with viscous developing jelly, an intensifying screen, and a positive image receiving sheet, has been developed primarily for medical radiography where the saving in time is a great boon. After exposure, the pack is passed through rollers which break the pad and spread the developing jelly. The dry positive print is ready in 1 min. Examples of the application of the method to industrial type examinations are given.—P. M. C.

Units Used In Industrial Radiography to Describe Strength of Cobalt-60 Sources. J. Kastner. (*Non Destructive Test.*, 1952, **11,** Summer, 21–23). The various methods employed to express the strengths of cobalt^{60} sources are the nominal disintegration rate or activity, the radium equivalent, the measured radiation output or ionization effect, and the effective or apparent distintegration rate. All are discussed with special reference to their advantages and disadvantages; the difficulty of comparison is emphasized. The radiation output estimated from the known disintegration rate is considered to be the most practical unit for workers in the radiographic field.—P. M. C.

Uses of Gamma Emitting Isotopes. W. S. Eastwood. (*Research*, 1952, **5,** Sept., 399–406). The uses of γ-ray emitting isotopes (generally prepared in a nuclear reactor) for industrial radiography for non-destructive testing, thickness measurement, radiation chemistry, and sterilization of food, are described. Health precautions are outlined, 20 γ-emitters are listed, and those with the most suitable properties are described in detail. (41 refcrences).—K. E. J.

Thermal Conductivity of Solids and Fluids from Studies of Aggregate Matter. H. S. Strickler. (*J. Chem. Phys.*, 1952, **20,** Aug., 1333–1334). A note on the relationship of the conductivity of fluids and porosity is presented. Experimental and calculated values for steel, lead, SiC and MgO are compared.—T. E. D.

Thermal Conductivity—Its Importance in the Behaviour of Heat-Resisting Materials. M. Pommellet. (*Métaux–Corrosion–Indust.*, 1952, **27,** July–Aug., 318–324). A review is made of the mathematics of heat transfer in connection with the components of engines running at very high temperatures.

Design Considerations for Heat and Corrosion-Resistant Castings. E. A. Schoefer. (*Machine Design*, 1952, **24,** Sept., 148–154). The room-temperature properties and applications of corrosion and heat-resistant cast alloys (ASTM Spec. No. A–296–49T and A–297–49T respectively) are tabulated and particular uses are outlined.—E. C. S.

Mechanical Strength of Carbon Steels at High Temperatures. M. A. Zaikov. (*Zhurnal Tehhnicheskoi Fiziki*, 1949, **19,** 6, 684–695). The influence of temperature, in the range 20° to 1300° C., on the mechanical strength of several carbon steels (C from 0·12 to 1·19%) has been investigated at elongation rates of 0·0004 and 0·04 in./sec. The results are presented in extensive tables and graphs, and qualitative correlations between the critical points and the carbon contents in the hot- and cold-working, precipitation-hardening, and recrystallization ranges of temperature are worked out.—R. A. R.

Heat-Resistant Alloys. R. A. Long. (*Machine Design*, 1952, **24,** Oct., 159–160). The author summarizes the uses of many heat-resisting alloys, drawing from his experience of the fabrication and development of materials for gas turbines, ram jets, and rockets.—E. C. S.

Creep-Tension Relations at Low Temperatures of Metals. J. D. Lubahn. (*Amer. Soc. Test. Mat. Preprint No.* 71, 1952). The most suitable method of determining creep behaviour from the results of tensile tests was to plot tension curves determined at various constant rates. This method cannot be used for metals such as copper or 61S–T (aged aluminium alloy) tested at room temperature as the stress-strain curve at different rates differs by less than the variations in deformation resistance among the specimens. The author suggests a new approach to this problem, based on a mathematical relationship which is derived from the hypothesis that mechanical behaviour is independent of past history. Preliminary practical results are in agreement with this hypothesis. The particular case of strain ageing is also considered and, by modifying the original relationship, the creep curve for any desired stress, using one measured creep curve and one measured tension curve, can be constructed.—B. G. B.

The Influence of Periodic Overstressing on the Creep Properties of Several Heat-Resistant Alloys. G. J. Guarnieri and L. A. Yerkovich. (*Amer. Soc. Test. Mat. Preprint No.* 72, 1952). Test equipment and procedures have been developed for conducting overload tests under conditions similar to those encountered in jet aircraft during operation. Overload tests were carried out on three steels, and it was possible to correlate the results into convenient design charts which relate the variables of stress, time, temperature, ratio of overstress to normal stress, and time percentages of overstress application. A method of calculating overstress characteristics has been successfully developed.—B. G. B.

A Survey of Embrittlement and Notch-Sensitivity of Heat-Resisting Steels. G. Sachs and W. F. Brown, jun. (*Amer. Soc. Test., Mat., Preprint No.* 74, 1952). The authors conclude, after examining published reports, that the steels investigated were found to be subject to embrittlement when heated under applied stress at 500° C. The magnitude of this effect varies with the alloy. The rate and magnitude of creep damage depends on the creep stress. The rate increases with increase in stress and the magnitude increases, with decrease in stress, to a maximum value. If creep damage exceeds a certain value, which depends upon the stress, it cannot be eliminated by subsequent heat-treatment. It has been

shown for one steel that the creep stress range which results in high creep damage is also the range which yields low rupture ductilities and a high notch sensitivity. Both creep damage and notch rupture sensitivity appear to be associated with a precipitation phenomenon.—B. G. B.

Effect of a Notch and of Hardness on the Rupture Strength of "Discaloy". F. C. Hull, E. K. Hann, and H. Scott. (*Amer. Soc. Test. Mat. Preprint No.* 75, 1952). Certain critical components of jet engines and gas turbines require materials of low notch sensitivity at service temperatures around 1200° F. In an Fe–Ni–Cr–Mo–Ti austenitic alloy (Discaloy) the effect of hardness, as varied by the hardener (titanium) has been related to the behaviour of the material in the presence of a notch. Creep-rupture and notched-bar rupture tests were conducted at 1000° and 1200° F. The plain bar rupture strength reached a maximum at a hardener content producing 2% rupture strain, and the notched-bar strength exceeded the plain bar strength in material in which the creep specimen had more than 5% rupture strain.

Report on ASTM Task Group for Determination of Elastic Constants. W. Ramberg. (*Amer. Soc. Test. Mat. Preprint No.* 158, 1952). Sixty-nine answers to a questionnaire, sent out concerning the measurement of the elastic constants, are considered. They show the need for better methods of determining the elastic constants of metals at elevated tempratures.—B. G. B.

The Influence of Temperature on the Elastic Constants of Some Commercial Steels. F. Garofalo, P. R. Melanock, and G. V. Smith. (*Amer. Soc. Test. Mat. Preprint. No.* 159, 1952). Elastic moduli in tension and in shear have been determined at various temperatures between 75° and 1500° F. for 21 different commercial steels, both plain and alloyed. With increasing temperature, both the tensile and shear moduli decreased until a certain temperature is reached beyond which a more rapid decrease was observed ; these temperatures were 700°, 900°, and 1300° F. for carbon, alloyed ferritic and austenitic stainless steels, respectively. Poisson's ratio did not change significantly with temperature.—B. G. B.

Influence of Sharp Notches on the Stress-Rupture Characteristics of Several Heat-Resisting Alloys. W. F. Brown, jun., M. H. Jones, and D. P. Newman. (*Amer. Soc. Test. Mat. Preprint No.* 76, 1952). Stress-rupture tests were carried out on a number of low-alloy steels, ferritic stainless steels, and austenitic alloys. Both unnotched and sharply notched bars were tested at various temperatures and the ductilities determined. All alloys tested were found to be subject to notch weakening, the time and temperature range of this effect and its magnitude varied with the alloy composition. The notch can apparently reduce the ductility of an otherwise ductile metal to the point where there is insufficient plastic flow to eliminate the initial stress concentration.—B. G. B.

Effect of Notch Geometry on Rupture Strength at Elevated Temperatures. E. A. Davis and M. J. Manjoine. (*Amer. Soc. Test. Mat. Preprint No.* 78, 1952). Three series of creep-rupture tests of notch bars with different notches were carried out at elevated temperatures on several heat-resistant alloys. The effect of stress level, grain size, hardness, ductility, and heat-treatment on notch sensitivity was also investigated. The effect of shank diameter can be virtually eleminated by using a 'standard' notch shape. It is postulated that fracture originates just below the surface at the root of the notch.—B. G. B.

Effect of Sigma on Strength and Ductility of 25 Cr, 20 Ni Steel. G. V. Smith and E. J. Dulis. (*Amer. Soc. Test. Mat. Preprint No.* 82, 1952). Samples of 25/20 Cr–Ni steels, initially cold-worked 35%, were heated at 1700°, 2000° and 2300° F. for 1 hr. and water quenched, then exposed for 7500 hr. at 1300° F. to promote precipitation of the sigma phase. These specimens were then compared with the unexposed metal. The sigma phase in this steel causes moderate strengthening at room temperature, severe loss in toughness even at temperatures as high as 500° F., and moderate loss of creep rupture strength at 1300° F.—B. G. B.

Recovery and Creep in an Alloy Steel. H. A. Lequear and J. D. Lubahn. (*Amer. Soc. Test. Mat. Preprint No.* 83, 1952). By interrupting a creep test and observing that the creep rate is higher after than before the interruption by amounts increasing with its duration, it is possible to establish that recovery (reduction of the amount of strain hardening) has occurred. Recovery occurs in quenched and tempered Cr–Mo–V steel at 1000° F., but not at 800° F. When recovery occurs, the plastic creep rate becomes constant—sooner for smaller stresses, and apparently immediately for a sufficiently small stress. When recovery does not occur, the plastic creep rate decreases continuously. Recovery does not cause the creep, however, for pronounced creep occurs at temperatures where there is no recovery.—B. G. B.

Exploratory Investigation of High-Temperature Sheet Materials. D. Preston. (*Amer. Soc. Test. Mat. Preprint No.* 85, 1952). This investigation was undertaken to determine the high-temperature mechanical properties of several materials commercially available in sheet form. The materials included three iron-base strain-hardening alloys, three iron-base age-hardening alloys and three cobalt-base alloys. Microstructures are discussed and creep, stress rupture and short-time tension and yield data are presented for these alloys from room temperature to 1800° F.—B. G. B.

The Time Laws of Creep. A. H. Cottrell. (*J. Mech. Phys. Solids*, 1952, **1,** Oct., 53–63). After a review of experimental and theoretical studies of transient creep, it is shown that logarithmic creep can be explained by the exhaustion theory. Wyatt's creep is explained in terms of the cutting of one dislocation through another as the unit process. The possibility of explaining Andrade's creep in terms of a modified exhaustion theory is indicated, but a solution is not likely to be found until work-hardening is introduced into the theory.

Creep of Metals. E. R. Parker. (*Amer. Soc. Met. Special Publication "High Temperature Properties of Metals,"* 1951, 1–40). This review includes a simple account of the theory of dislocations, the stages of creep and various theoretical equations for strain against time. Creep is related to metallurgical factors, metal composition, and the environment. Creep-resistant metals are briefly discussed. (84 references).

Stress Rupture Testing. N. J. Grant. (*Amer. Soc. Met. Special publication, "High Temperature Properties of Metals,"* 1951, 41–72). The superiority of the stress-rupture test over tensile testing at high temperatures is emphasized, and practical details are given of the test procedure and method of plotting the results. The concept of equicohesion is outlined. (24 references).—E. T. L.

High Temperature Fatigue Testing. H. J. Grover and H. C. Cross. (*Amer. Soc. Met. Special Publication, "High Temperature Properties of Metals,"* 1951, 73–92). The author deals with the following factors which are important in rotating-bend testing, repeated bend testing, and axial load testing : Temperature, loading, alignment of the specimen, measurement of the stress and deformation, and the choice of a failure criterion.—E. T. L.

Some Experiences in Service (Power, Oil and Chemical Plants). J. J. B. Rutherford. (*Amer. Soc. Met. Special Publication, "High Temperature Properties of Metals,"* 1951, 133–170). Practical experiences of the effects of scale, carburization, intergranular corrosion, carbide precipitation, salt attack, pitting, attack by vanadium pentoxide, transformation of steel to ferrite and graphite, embrittlement, thermal stresses, and forms of fatigue are reported.—E. T. L.

Rheology of Metals at Elevated Temperatures. A. E. Johnson and N. E. Frost. (*J. Mech. Phys. Solids*, 1952, **1,** Oct., 37–52). An investigation of the general stress, time, and temperature dependence of the creep, plastic strain, and relaxation properties of a low carbon steel, an RR59 type aluminium alloy, and a 2% Al magnesium alloy, in temperature ranges of practical engineering interest, is described. The investigation is not complete ; results to date are examined in the light of existing theories.—J. G. W.

Creep Strength of Low-Alloy Chromium-Molybdenum Steels. W. Tomaszczyk and Z. Borysowski. (*Prace Instytutu Metalurgii*, 1952, **4,** 3, 261–271). [In Polish]. The results of investigations of creep strength of low-alloyed Cr–Mo steels, including the effect of deoxidation, chemical composition and heat-treatment are discussed. The minimum chromium addition to obtain stability of these steels against graphitization has been determined. A table of their properties over the temperature range at which they are used, is given.—V. G.

A Comparison Between Fe-Cr-Al and Ni-Cr Alloys for High Temperature Service. G. Hildebrand. (*Proc. First World Metallurgical Congress*, Sept., 1951, 632–636 : Amer. Soc. Met. 1952).

Are our Structural Materials for Use in Superheated Steam Endangered by Graphitization ? C. Holzhauer. (*B.W.K.*, 1952, **4,** Apr., 134–135). The appearance of grephite particles in structural steel in use at high temperatures was observed

and examined in the U.S.A. and in Germany. The theory relating to the conditions producing graphitization are considered and some practical conclusions are derived from this.

Determination of Mechanical Properties of Carbon Steel at High Temperatures. V. Montalvo Montero. (*Anal. Mec. Elect.*, 1951, **27**, Nov.–Dec., 314–321). [In Spanish]. The effect of temperature on the mechanical properties of steel and methods of measuring creep are discussed.—R. S.

Investigation of Materials for Steam Turbine Manufacture. W. Stauffer. (*Escher-Wyss News*, 1950–51, **23–24**, 91–97). All materials used at Escher Wyss are tested to verify that they have the required properties. Normal modern testing techniques are employed. Some interesting tests performed on materials used in steam turbine manufacture are reported. They include creep tests of long duration on Cr–Mo steel and Mo steel, metallographic investigations on austenitic corrosion-resistant steel containing tungsten and titanium, X-ray course structure and γ-ray tests, non-destructive testing by the magnetic powder method, dynamic fatigue tests, water jet impact tests, the testing of synthetic carbon for labyrinth seals, and spectrographic investigations.—T. E. D.

Mathematical Laws and Experimental Data Concerning Creep of Heat-Resisting Alloys for Use In Gas Turbines. G. Vidal and A. Loupoff. (*Métaux–Corrosion–Indust.*, 1952, **27**, Sept., 358–370). A mathematical examination is made of published data concerning the failure of heat-resisting alloys at elevated temperatures. The authors show that, whilst great care is needed in the extrapolation of practical results to different conditions, it is possible to predict mathematically the number of tests required to evaluate the properties of alloys operating at high temperatures.—B. G. B.

Introduction to the Theoretical Determination of Annealing Conditions for Relaxation of Stresses in Heat-Resisting Steels. G. Vidal and A. Loupoff. (*Métaux–Corrosion–Indust.*, 1952, **27**, Sept., 371–378). A mathematical study of the stresses occurring in steels operating at high temperatures is made. Whilst the mathematical theory is not complete, a number of predictions can be made.—B. G. B.

Creep Relationships and Their Application to Pipes, Tubes, and Cylindrical Parts under Internal Pressure. R. W. Bailey. (*Proc. Inst. Mech. Eng.*, 1951, **164**, No. 4, 425–431). The author develops the mathematical analysis and theory of the following : Creep strain relations for simple and complex stress at any stage (primary, secondary, or tertiary), of creep ; stress distribution and creep of cylindrical parts under internal pressure, and related formulae for wall thickness ; determination of design temperature to provide for specific temperature and pressure fluctuations ; action of creep in removing the thermal expansion loading of pipeline.—P. M. C.

Effect of Hydrogen on the Deformation and Fracture of Iron and Steel in Simple Tension. P. Bastien and P. Azou. (*Proc. First World Metallurgical Congress*, Sept,, 1951, 535–552 : Amer. Soc. Met., 1952). A study is presented which attempts to establish the exact effects of hydrogen on the deformation and cohesive strength of iron and mild steel in simple tension. Some conclusions are drawn with regard to the probable mechanism of hydrogen embrittlement.

Analysis of Process of Absorption of Gases by Metals. II—Absorption of Hydrogen and Nitrogen by Iron. A. D'yakonov and A. Samarin. (*Izvestia Akademii Nauk S.S.S.R., Otdelenie Tekhnicheskikh Nauk*, 1946, Jan., 121–126). Equations for the solubility of hydrogen and nitrogen in α-, γ-, δ- and liquid iron are presented, and the heat contents and free energy changes accompanying the absorption of these gases by iron are calculated.—R. A. R.

The Thermal Accommodation Coefficient of Gases and Their Adsorption on Iron, A. E. J. Eggleton, and F. C. Tompkins. (*Trans. Faraday Soc.*, 1952, **48**, Aug., 738–749). Advantage has been taken of the large difference in the thermal accommodation coefficient of neon at the surface of pure iron wire in an investigation of the methods of producing a surface free from contaminants. Hydrogen reduction methods followed by flashing *in vacuo* at 1200° C. proved ineffective, but neon-ion bombardment was successful if this was preceded by a preliminary treatment with nitrogen ions. Values of the accommodation coefficient of neon at surfaces covered with hydrogen, oxygen, and nitrogen have been measured. The significance of these results from the viewpoint of chemisorption phenomena is discussed.—J. C. B.

The Generation of Gas Bubbles at the Shrinkage Boundaries of Built-Up Crankshafts for Diesel Engines. S. Amari and E. Ando. (*Trans. Inst. Marine Eng.*, 1952, **64**, Sept., 185–200). The results of a number of tests on diesel engine crankshafts manufactured in Japan are given. The authors show that the gas bubbles generated at the shrinkage boundaries contain hydrogen. This bubbling phenomena was not the cause of slipping at the shrinkage surface—the slip was due to lubricant on the shrink fit surface. No defect such as hair-line cracks were discovered and the shafts were mechanically satisfactory. In Japan, a single heat-treatment is carried out after finish-forging, whereas in England a further heat-treatment is given after rough-machining, and it is suggested that a large proportion of the hydrogen is evolved during this second treatment, so that the emission of hydrogen after assembly is very small.—B. G. B.

The Permeability of Steels to Lithium. A. Hérold, P. Muller, and P. Albrecht. (*Compt. Rend.*, 1952, **235**, Sept. 29, 658–659). Lithium cannot pass through pure iron below 1100° C. but does so through steels by reaction with other elements, notably carbon, thus creating fissures. Prolonged heating in hydrogen renders the steel impermeable.—A. G.

Possibilities of Saving Molybdenum in Special Steels. R. Štefec. (*Hutnické Listy*, 1952, **7**, 8, 395–398 ; 9, 479). [In Czech]. Economic reasons for finding substitutes for molybdenum are discussed and the effects of molybdenum in steels are examined. The composition of a number of Czechoslovak high-speed, tool, and structural steels for cementation and carburizing as well as steels of high creep-resistance, in which molybdenum had been replaced by other elements, are given, with data on their properties.—P. F.

Considerations on the Substitution of Tungsten for Molybdenum in Structural Steels. P. Gomez Baeza. (*Tecnica Metalurgica*, 1952, **8**, Feb., 45–61). [In Spanish]. The author discusses Spanish production of wolfram and the Fe–Mo, Fe–W, Mo–C, and W–C systems. Molybdenum and tungsten are considered to be the carbide-forming elements in the Fe–C–Mo and the Fe–C–W systems. The influence of molybdenum and tungsten on (*a*) hardenability, (*b*) the critical points of Fe–C alloys, (*c*) the sub-critical transformations of these alloys, (*d*) the mechanical properties of low alloy steels, and (*e*) temper brittleness is discussed. It is evident that molybdenum can be replaced by tungsten in low alloy steels.—R. S.

Effect of Austenitic Grain Size on Nickel-Chromium Steels. A. B. Chatterjea and B. R. Nijhawan. (*J. Sci. Indust. Res.*, 1952, **11B**, Sept., 388–391). The addition of about 0·05% aluminium to Ni–Cr alloy steels refines the austenitic grain size, and improves the impact strength. The susceptibility to temper-embrittlement is also reduced.—E. T. L.

Nickel and Nickel-Base Alloys. H. O. Teeple. (*Indust. Eng. Chem.*, 1952, **44**, Oct., 2325–2338). Developments in the knowledge and use of materials containing 40% or more of nickel, or appreciable quantities of cobalt, during 1951–52 are surveyed. (293 references).—K. E. J.

Influence of Nickel upon Quantitative Characteristics of the Coalescence of Carbides in Isothermal Tempering. S. Z. Bokshtein. (*Zhurnal Tekhnicheskoi Fiziki*, 1950, **20**, 3, 327–333). The influence of nickel on the size of carbide particles in steel after quenching and isothermal tempering was investigated. Nickel accelerated the coalescence of carbides and increased their particle size, which is explained by an increase in the rate of diffusion of carbon in ferrite.—V. G.

The IVth International Congress on Machine Manufacture—Boron Steel. (*Usine Nouvelle*, 1952, Oct. 2, 49–50 ; Oct. 9, 47–51). Work on boron steels in America and more recently in Germany is surveyed. The influence of boron on deoxidation and denitriding is outlined. Its effect on steels during heat-treatment and changes in mechanical properties, and its influence on cementation are described.—T. E. D.

Effect of Boron on the Mechanical Properties of Low-Alloy Steels. R. Wilcock. *J. Iron Steel Inst.*, 1953, **173**, Apr., 406–419). [This issue].

The Influence of Boron on Case Hardenability in Alloy Carburizing Steels. C. F. Jatczak and E. S. Rowland. (*Amer. Soc. Met., Preprint No.* 14, 1952). The end-quenched test on carburized specimens was used to evaluate the effect of boron on the case-hardenability of four steels, with between 0·60 and 1·00% carbon, with different heat-treatments. Tests were also made with similarly treated steels, identical except for the absence of boron. The effect of boron on the case-hardenability decreased with increase of carbon content.

Boron Steels. H. B. Knowlton. (*Machine Design*,1952, **24,** Oct., 133–136). The presented status of boron-treated steels is reported on favourably.—R. A. R.

Effect of Aluminium on the Properties of Low Carbon Steels. K. Born and W. Koch. (*Stahl u. Eisen*, 1952, **72,** Oct. 9, 1268–1277 : *Iron Steel Inst. Translation Series No.* 463, 1953). The effect of small contents of aluminium in low carbon steels on grain size, susceptibility to ageing, and yield point have been studied as a function of the state of combination of the aluminium. Results on melts from a 25-lb. H.F., a 5-ton arc, and a 90-ton open-hearth furnace indicate that aluminium nitride is the controlling influence on structure and mechanical properties. This nitride reduces the γ-grain size, makes the steel insusceptible to ageing, and raises the yield point. Quenching experiments have shown that the nitride changes at 1000° C.; it can no longer be isolated and is apparently dissolved in the austenite. This explains the resistance of fine-grained steels to overheating at this temperature, and the grain growth that takes place. No direct connection could be found between the properties of steel and the precipitated alumina, although, as electron microscope studies have shown, it is present in particles less than 0·05μ in dia.—J. P.

The Effect of Small Aluminium Additions on Alloy Steel. M. Signora. (*Proc. First World Metallurgical Congress*, Sept., 1951, 260–267 : Amer. Soc. Met., 1952). The effect of additions of aluminium in nickel-chromium structural steel was examined. The most pronounced effect found was that aluminium refines the grain, independently of pouring temperature.—D. L. C. P.

The Importance of Submicroscopic Inclusions in the Development of High-Strength Heavy Structural Steels. F. Nehl. (*Stahl u. Eisen*, 1952, **72,** Oct. 9, 1261–1267). The necessity for increased yield point in manganese-silicon weldable steels has led to systematic investigations into the manganese content required for the highest yield-stress/tensile-strength ratio. Microexaminations and tensile tests on specimens of weldable Mn-Si structural steels, quenched from 870° C. to various temperatures, have shown that an intermediate stage structure without martensite inclusions gives the largest increase in yield stress and yield/tensile strength ratio. In all probability, submicroscopic inclusions of critical size are the cause of this. The effect is further increased by annealing at 600° C., during which there is a decomposition of the intermediate structure. The main cause of the high ratio in aluminium-killed weldable Mn–Si steels containing up to 1·2% Mn is aluminium nitride inclusions of critical size. These are precipitated, however, only under certain definite conditions. Investigations aimed at improving the hot strength of copper-nickel steels by precipitation of critical sized carbide inclusions indicated that additions of 0·2–0·3% of molybdenum or vanadium or less than 0·03% (tantalum + niobium) or titanium are sufficient to improve the ratio of the yield stress at 350° C. to tensile strength at ordinary temperatures, provided that the steels are annealed below 650° C. after normalizing. Tests of over 10,000 hr. at 350° C. indicate no change in the unusually high yield point and also no embrittlement.—J. P.

High-Tensile Steels. C. H. Stevenson. (*Aircraft Prod.*, 1952, **14,** Dec., 411). The use in aircraft construction of steels with a tensile strength of up to 220,000 lb./sq. in., and with good ductility, is discussed. The properties of ' Hy-Tuf ' alloy, made by the Crucible Steel Co. of America, are given, and techniques for machining it are outlined.—T. E. D.

Progress in Stainless and Heat Resisting Materials. J. Lomas. (*Eng. Boiler House Rev.*, 1952, **67,** Nov., 325–326). A review is presented of recent developments concerning stainless and heat-resisting steels of particular interest to the power engineer.—D. L. C. P.

Stainless Steels and other Ferrous Alloys. W. A. Luce. (*Indust. Eng. Chem.*, 1952, **44,** Oct., 2346–2359). Developments in stainless steels throughout the world during 1951–52 are reviewed. (287 references).—K. E. J.

Converter Steel Blown with Pure Oxygen, and Its Properties. W. Kühnelt. (*Stahl u. Eisen*, 1952, **72,** Aug. 14, 1004–1010). The properties of rimmed mild steel produced by top-blowing with oxygen have been investigated. The tensile strength, cleanliness, segregation behaviour, the hot and cold deformability, deep-drawing properties, and impact strength in the aged and non-aged conditions show that it is at least equal to rimmed open-hearth steel. The mechanical (including fatigue) properties of killed oxygen-blown converter steel in the rolled or normalized condition are quite as good as those of open-hearth steel of the same composition. End-quench tests on killed oxygen-blown steel, as well as heat-treatment and case-hardening trials, have proved the material to be suitable for heat-treating and carburizing. Even the cleanliness necessary for high quality heat-treating and case-hardening steels can be readily maintained. Rails made from this steel have performed successfully under impact, bend and fatigue tests.

Properties of Steel Refined in the Converter by Top-Blowing with Pure Oxygen. H. Hauttmann. (*Stahl u. Eisen*, 1952, **72,** Aug. 14, 1011–1018). Investigation of the properties of rimming, killed and low-alloy steels made by refining pig iron with pure oxygen has indicated that these are high-grade materials equal, and in many respects superior, to open-hearth steels. It is possible, by the new process, to produce rimming steels with increased resistance to ageing in thicknesses up to 30 mm. without resorting to high cooling rates. The ductility of rimming and aluminium-killed steels is particularly marked. These new steels are characterized by a high yield point and, due to their cleanliness, by good weldability. They are suitable for cold deformation operations and in this connection are superior to ordinary open-hearth steels. Low-carbon rimmed steel is made more easily by the new process than in the open-hearth furnace. There is nothing to prevent oxygen-refined steel from being used for ship and boiler plate. In corrosion resistance, electrical resistance, and hardening on welding, the new material is comparable with open-hearth steel. A heat for silicon-iron transformer sheet has been made satisfactorily.—J. P.

Japanese Substitution Steels in the Last War. S. Nishikiori. (*Proc. First World Metallurgical Congress*, Sept. 1951, 483–489 : Amer. Soc. Met., 1952). The author tabulates the types of substitute steels used during the 1939–45 war. As nickel, molybdenum and tungsten were extremely scarce, steels based on silicon, manganese, and chromium with suitable carbon contents and heat-treatments had to be used. Although inferior, these steels were successfully employed. Suggestions for improvement are made from the experience gained.—D. L. C. P.

Basic Electric Arc Steel *Versus* Acid Open Hearth Steel for Roller Bearings. B. Kjerrman. (*Proc. First World Metallurgical Congress*, Sept., 1951, 502–505 : Amer. Soc. Met., 1952). The author has investigated the failure of roller and ball bearings. The number, size, and shape of non-metallic inclusions, and the presence of dissolved gas in the steel are important. Open-hearth steel was found to have a lower nitrogen content than that from electric arc furnaces, and bearings made from it gave better performance under laboratory conditions.—D. L. C. P.

The Mechanical Properties of Iron and Some Iron Alloys of High Purity. W. P. Rees. (*Proc. First World Metallurgical Congress*, Sept., 1951, 506–534 : Amer. Soc. Met., 1952). The transition temperatures of high-purity iron to which controlled amounts of carbon and manganese were added, were studied with Charpy impact and Izod **V**-notch tests.—D. L. C. P.

Mechanical Properties of Spring Steel. T. Mitsuhasi, M. Ueno, R. Nakagawa and K. Tsuya. (*Proc. First World Metallurgical Congress*, Sept., 1951, 573–579 : Amer. Soc. Met., 1952). The authors report on research to determine the optimum heat-treatment for a 1% carbon spring steel for watches and clocks.—D. L. C. P.

Contribution to the Study of the Brittleness of Hardened and Tempered Steels Between 250° and 400° C. P. Gomez Baeza and J. A. Garcia Poggio. (*Inst. Hierro Acero*, 1952, **5,** Apr.–June, 614–627). [In Spanish]. The impact strength of Cr–Mo and Cr–Ni–Mo steels, after various heat-treatments, has been studied at temperatures down to −70° C.—R. A. R.

Normality and Abnormality of Steels. S. Feliu Matas. (*Rev. Ciencia Apl.*, 1952, **6,** May–June, 229–234). [In Spanish]. Factors affecting the final structure of steel are examined. The effects of various types of carburizing agent on the normal and abnormal behaviour of steel are discussed. Some cases are mentioned where the hardness of the hardened and tempered surface indicates normal behaviour of the steel, whilst the McQuaid-Ehn test indicates abnormality.—R. S.

The Influence of Graphite and of the Matrix on Some Mechanical Properties of Ordinary Cast Irons. I. W. Grundig. (*Bol. Assoc. Brasil. Metais*, 1952, **8,** Jan., 77–107). [In Portuguese). The author studies the effect of graphite and of the matrix on the strength and workability as determined by the tensile, bend, and hardness tests in predominantly pearlitic and hypereutectoid cast irons.—R. S.

Experiences when Austempering an 18/4/1 W-Cr-V High-Speed Steel. F. S. Ayma. (*Inst. Hierro Acero*, 1952, **5**, July, 730–740). [In Spanish]. An investigation of an 18/4/1 W–Cr–V high-speed steel, in which the physical properties after ordinary annealing and isothermal heat-treatment were determined, is reported.—R. A. R.

Specification for the Acceptance of Metals to be Used in the Construction of Pressure Vessels Subject to the Control of National Association for Combustion Control (N.A.C.C.). (*Calore*, 1952, **23**, Sept., 460–471). The specification covers ferrous and non-ferrous metals, and in particular deals with carbon steels, hot rolled steels, forged steel for use in gas or steam pressure vessels, cast steels, stainless and alloy steels, cast irons, rolled copper, and aluminium and aluminium alloys. The tolerances and methods of testing are specified.

The Standardization of Spring Steel Wires. E. Siebel and W. Panknin. (*Stahl u. Eisen*, 1952, **72**, Sept., 25, 1193–1195). Investigations have shown that the present standard DIN 2076 for spring steel wires is not satisfactory since it gives no details of either the deformability of the material or its purity. In revising this specification, the question of deformation characteristics must be considered. According to the present investigations, the reduction of cross section during tensile testing, the amount of twist to fracture in the torsion test, and an evaluation of the torsion specimen with respect to surface quality, uniformity of twisting and appearance of the fracture, may make possible a satisfactory assessment of the deformability of spring wires.—J. P.

High-Strength Weldable Structural Steels: Possibilities for Further Improvement. F. Nehl. (*Iron Coal Trades Rev.*, 1952, **165**, Sept. 5, 545–546). The author discusses the necessary requirements of steels suitable for welded structures and describes recent German developments in this field. Particular attention is given to a new low-alloy steel (0·20% C, 0·45% Si, 0·95% Mn) of high weldability with a tensile strength of 31–38 tons/sq. in.—G. F.

The Development of Spheroidal Graphite Cast Iron. W. W. Braidwood. (*Fonderia Ital.*, 1952, **1**, Sept., 154–160). The author reviews the development of spheroidal graphite cast irons and shows how, once the initial difficulties had been overcome, techniques made rapid progress. Methods adopted to neutralize harmful elements are referred to, and details are given of the use of metallic magnesium. The author examines the mechanical, heat-treatment, and other properties of spheroidal graphite iron as well as what are termed 'special spheroidal irons'. (14 references).—M. D. J. B.

The Situation in France Regarding Mechanical Testing for the Acceptance of Grey Iron Castings. E. Doat. (*Fonderie*, 1952, Mar., 2846–2880). The author recalls the circumstances preceding the choice of various tests of the mechanical properties of grey cast iron. He describes the tensile test and the to be obtained for different sizes of test bar.—R. S.

The Cast Irons of the Nazionali Cogne and Their Uses in Foundry Work. G. Somigli. (*Fonderia Ital.*, 1952, **1**, Sept., 137–148). A brief description is given of the manufacture of hematite irons at Cogne. Details are given of the types manufactured and their analyses, properties, and structure. Fremont test (shear and static bend), and specifies the values It is shown that these characteristics, particularly those of the high grade irons, depend upon the cast characteristics and burdens. Details are given of the special irons manufactured at Cogne for subsequent transformation into spheroidal graphite irons.—M. D. J. B.

The Correlation Between Analysis and Certain Mechanical Properties of Ordinary Grey Cast Iron. J. Drachmann. (*Gjuteriet*, 1952, **42**, July, 108–114). [In Swedish]. Results are presented of a statistical evaluation of a number of routine casting analyses show that it is possible, without using expensive equipment, to examine the influence of carbon, silicon, sulphur and MnO on machinability, Brinell hardness, and chill depth of grey cast iron, the example chosen containing 0·60% P. The six possible binary correlation diagrams for the variables involved, machinability (or hardness or chill depth), carbon, silicon, and MnO are calculated, the six associated correlation coefficients are evaluated, and the lines of regression ascertained by the Frisch graphical method. Quadratic approximations were used in the case of hardness and chill depth. Results show the latter to be at a minimum with 2·70% Si, whilst the other two qualities were dependent on the variations of carbon and silicon with the carbon equivalent.—G. G. K.

Present State of German Testing of Grey Iron Castings. A. Wittmoser and W. Seeliger. (*Giesserei*, 1952, **39**, Sept. 18, 477–483; Oct. 2, 533–538). Following a resolution by the International Committee for the Testing of Castings, at the Brussels Conference in 1951, the authors undertook a critical survey of German testing procedures. The tests comprise the measurement of mechanical, physical, chemical, and metallurgical properties, some of which are described in detail. Far-reaching standardization of tests is needed, and detailed recommendations are made.—J. G. W.

How Statistical Quality Control Can Help You. D. H. W. Allan. (*Canad. Metals*, 1952, **15**, Sept., 48–50). Statistical quality control is defined as statistical analysis of inspection data, process control through control charts, and the use of sampling plans to insure, within set limits. the quality level of material passing through an inspection station. Its greatest saving is the reduction of scrap and repetitive work.—J. C. B.

METALLOGRAPHY

Some Recent Developments in Metallurgical Microscopy. R. C. Gifkins. (Australian Inst. Metals: *Australasian Eng.*, 1952, May 7, 63–70). Brief details are given of all the accepted methods for the examination of (*a*) surface levels and colouration, and (*b*) surface orientations and movement in the plane of the surface. The former group includes normal incidence and oblique illumination, phase-contrast, stopped down diffraction effect, stop-contrast, dark ground, interference fringes, and the reflecting microscope. The second group comprises etch pits, the movement of surface markings, and applications of polarized light. (30 references).

Direct Examination of Solid Surfaces Using a Commercial Electron Microscope in Reflection. J. W. Menter. (*J. Inst. Metals*, 1952, **20**, Nov., 163–167). The modifications necessary for the conversion of the Metropolitan-Vickers EM3 electron microscope for use in the direct examination of solid surfaces by reflection, are described. The geometry, resolution, and the depth of focus of the image are discussed. A range of specimens with widely varying surface topography has been examined in order to illustrate the value of the technique and to determine the effect on the image of varying the angle of inclination of the specimen.—J. C. B.

Quantitative Metallography with the Electron Miscroscope. A. L. Ellis and F. K. Iverson. (*Amer. Soc. Test. Mat. Bull.*, 1952, July, 66–69). The principles of quantitative metallography are presented and the special problems involved in the application of the electron microscope are discussed. It is shown that the electron microscope can be used advantageously when the particle size is very small. Representative sampling and reproducibility are discussed, and the results obtained by this and other methods are compared.—B. G. B.

Corpuscular Microscopes. R. Simonet. (*Usine Nouvelle*, 1952, Oct. 16, 29–30, 47). The mode of operation of the electron microscope is described, and the various types are listed. Electric and magnetic lenses are mentioned. Applications of the instrument are outlined, and some micrographs are shown. Finally, the proton microscope is briefly described. This instrument is capable of a resolution of 3Å against 50Å with the best electron microscope. This improvement is possible by eliminating most of the spherical aberration and diffraction.—T. E. D.

Electron-Diffraction Study of Iron Carbides in Bainite and Tempered Martensite. A. E. Austin and C. M. Schwartz. (*Amer. Soc. Test. Mat. Preprint No.* 165, 1952). Samples of bainite and tempered martensite, etched to leave the carbide phases in relief, have been examined by electron diffraction. A plain carbon steel of eutectoid composition was used. Cementite was found in samples isothermally transformed, or quenched and tempered at or above 500° F. Both cementite and epsilon iron carbide exist at 500° F. In martensite tempered at 400° F. only epsilon iron carbide was found.

An Electron Microscope Study of the Development of Fatigue Failures. W. J. Craig. (*Amer. Soc. Test. Mat. Preprint, No.* 167, 1952). A pictorial comparison is made of deformation marks on the surface of alpha brass, ingot iron and aluminium under both static and repeated loading. Under static loading the deformation is a general process taking place in all crystals, while under conditions of repeated loading the deformation is an extremely localized phenomenon taking place in a few crystals or portions of crystals.—B. G. B.

Electron Microstructure of Bainite in Steel. (*Amer. Soc. Test. Mat. Bull.*, 1952, May, 62–65). The majority of the carbides found in lower bainite are extremely small plate-like particles arranged in parallel array within long narrow ferrite needles. The small carbide platelets appear as cross-striations at an angle of 55° to the ferrite needle axis. The transition from lower to upper bainite occurs in the transformation temperature range between 600 and 650° F. This transition is characterized by two distinct changes in the microstructure. First, the size, shape, and orientation of the carbide particles change and then the needles of ferrite, which account for the marked acicular nature of lower bainite, disappear. As the temperature of transformation is further increased above the transition range, the size of the individual upper bainite areas increases and the carbide stringers become larger and longer. Electron micrographs of specimens partially transformed to upper bainite at 850° F. show the structure to consist of stringers of carbide surrounded by a sheath of ferrite. Similarly, 'feathery bainite' is a parallel array of carbide needles or plates, in a ferrite matrix. Small amounts of fine pearlite can co-exist with upper bainite in this steel at transformation temperatures as low as 850° F. Although preliminary electron diffraction indicates the presence of some hexagonal ϵ-iron-carbide in bainite transformed at 500° F., the results suggest that the carbide platelets are cementite in the cross-striated needle structure in 550° and 600° F. lower bainite.—B. G. B.

Electron Tubes for Industry and Research. C. C. Gee. (*Electronic Eng.*, 1952, **24,** Nov., 540–544). The various types of electron tubes are classified, according to their modes of releasing full electrons, into the following categories: Thermionic valves, cathode-ray tubes, arc-action gas-filled tubes, X-ray tubes, photo-electric devices, and cold-cathode tubes. These are sub-classified, and a very comprehensive table is given, with the application of each type.—L. H.

Moving Pictures of Metal Structures Develop New Technique for Metal Structure Study. R. B. Pond and N. K. Chen. (*Iron Age*, 1952, **170,** Aug. 7, 122–126). Motion pictures of the solidification of low melting point alloys have been taken through a microscope. This method, combined with a special micro-tensile testing machine equipped with strain gauges, has been used for the study of the plastic deformation of high-purity aluminium crystals. It is suggested that this method could be used to supply many of the missing links between single crystal plasticity and the plasticity of polycrystalline aggregates.—A. M. F.

Time Magnified High Speed Photography: Steel Research Tool. R. A. Buchanan. (*Steel*, 1952, **131,** July 7, 90–91). The use of a 3500 frames/sec. cinecamera at the United States Steel Company's research laboratory is described. At this speed, lighting is one of the biggest problems.—A. M. F.

The Functions of the Government Metallurgical Laboratory. L. Taverner. (*J. Chem. Met. Min. Soc. S. Africa*, 1952, **52,** Apr., 225–240). The objects and aims of the Government Metallurgical Laboratory, Johannesburg, are discussed. The history of the Laboratory is given in some detail, and the apparatus available and work carried out in each department are described.—D. H.

Perchloric Acid—It Can Be Used Safely. W. B. Sobers. (*Foundry*, 1952, **80,** Aug., 95, 216). This is a description of a fuming hood fitted with water sprays, with some comments on the storage and handling of perchloric acid in metallurgical laboratories.—E. T. L.

Metallographic Analysis of Ferro-Alloys and Alloy Steels with the Aid of Heat Tinting. S. S. Gorelik and B. G. Livshits. (*Zavodskaya Laboratoriya*, 1950, No. 5, 578–579). [In Russian]. The application of temper colours to the identification of phases in ferro-alloys and alloy steels is described. The preliminarily polished specimen is heated, repeatedly if necessary, for up to 90 sec. For stainless and high alloy steels and ferro-alloys, the heating temperature is 700–750° C., for lower alloy-steels, 600–650° C. Annotated photomicrographs illustrate the appearance of many phases, including: Iron tungstide, tungsten carbide, eutectic of tungsten carbide and alloyed α-solution, α-solution for ferro-tungsten; and the inter-metallic compound FeV, α-solution.—S. K.

The Electrolytic Extraction of Carbides in a Rapidly Quenched Steel. J. Papier. (*Compt. Rend.*, 1952, **235,** Sept. 22, 619–621). The technique is described using 5% hydrochloric acid and a c.d. of 1 amp./sq. dm. Results on a high-speed steel containing tungsten, chromium, and vanadium are given. The carbide contains less chromium and more vanadium than the matrix and corresponds mainly to a compound of the type M_6C.—A. G.

The Existence of a New Phase in Chromium-Molybdenum Steels. L. Habraken. (*Rev. Univ. Mines*, 1952, Series 9, **8,** July, 281–284). The results of electron microscope and X-ray diffraction studies are given. The author shows the existence of a new phase in chromium-molybdenum steels. This phase is considered to be responsible for the creep resistance of these steels at high temperatures (550° C.)—B. G. B.

Do Metals Recrystallize? P. A. Beck. (*Trans. Amer. Inst. Min. Met. Eng.*, **194,** *J. Met.*, 1952, **4,** Sept., 979–980). The author discusses the traditional concept of recrystallization which takes place in the annealing of cold-worked metals. Recent evidence indicates that this traditional concept is in error, and that recrystallization is fundamentally akin to a type of grain growth occurring in a substantially strain-free matrix with strong single orientation texture.—G. F.

Metals and Alloys Are Made Up of Crystals. (*J. d'Informations techniques des Industries de la Fonderie*, 1952, June, 19–22). A brief account of the different crystal structures occurring in metals and alloys is given.—B. G. B.

Recrystallization Kinetics of Low Carbon Steel. S. F. Reiter. (*Trans. Amer. Inst. Min. Met. Eng.*, **194,** *J. Met.*, 1952, **4,** Sept., 972–979). Isothermal recrystallization curves of 0·08% C. steel at 700°, 670°, and 640° C after 8% and 9% elongation, and for 0·15% C steel at 700° and 670° C. after 9% elongation, are presented. The effects of deformation, temperature, and ageing on nucleation and growth rates are discussed, and the activation energies for nucleation, growth, recrystallization, and grain boundary migration are calculated.

Observations on Growth and Etch Phenomena on Haematite (Fe_2O_3) Crystals. A. R. Verma. (*Proc. Phys. Soc.*, 1952, **65B,** Oct. 1, 806–811). New observations of growth and etch features on hematite crystals are reported. The simplest growth patterns are the elementary 'molecular growth spirals': the step height between successive arms of these is shown to be equal to the X-ray value of the repeat distance. The etch figures are also of molecular thicknesses, consisting of a very large number of similarly orientated small triangular figures, and other shapes. (10 references).—K. E. J.

Crystal Orientation in Cold-Rolled Silicon-Iron Sheet. I. Gokyu and H. Abe. (*Proc. First World Metallurgical Congress*, Sept., 1951, 580–590: Amer. Soc. Met., 1952). This study presents some experimental results on the rotation of crystal orientation in a silicon-iron sheet during cold-rolling and annealing.—D. L. C. P.

The Orientation Relations between Cementite and α-Iron. J. J. Trillat and S. Oketani. (*Acta Crystallographica*, 1952, **5,** July, 469–471). [In French]. The preparation of single crystal films of α-Iron, which are then cementized at 500° C. in $CO + H_2$, is described. The surface was examined by electron diffraction and the results are given. The interpretation of the patterns is explained, and it is shown that prolonging the reaction leads to a progressive disorientation of the cementite.

Solid State Physics in Electronics and in Metallurgy. W. Shockley. (*Trans. Amer. Inst. Min. Met. Eng.*, **194;** *J. Met.*, 1952, **4,** Aug., 829–842). From experience in two fields of solid state physics, namely, transistor electronics and dislocation theory, the author shows how metallurgical industry will benefit from supporting fundamental research on dislocations. The two fields and their relationship are discussed, and recent results are given which illustrate the application of dislocation theory, the examples chosen being concerned with growth spirals, grain boundary energies, the strength of small metal crystals, and the surface hardening of aluminium crystals.

Self-Diffusion in Alpha Iron under Uniaxial Compressive Stress. F. S. Buffington and M. Cohen. (*Trans. Amer. Inst. Min. Met. Eng.*, **194;** *J. Met.*, 1952, **4,** Aug., 859–860). A research programme has been undertaken to determine the rate of self-diffusion in high-purity iron, with the specimens subjected to stress during the course of diffusion, and results are reported for body-centred cubic iron at a diffusion temperature of 890° C. The diffusivity is proportional to the strain rate and not to the strain itself.—G. F.

One-Dimensional Diffusion with the Diffusion Coefficient a Linear Function of Concentration. R. H. Stokes. (*Trans. Faraday Soc.*, 1952, **48,** Oct., 887–892). The differential equation of diffusion, with a diffusion coefficient which is a linear function of concentration, is discussed for the case of a one-dimensional diffusion from an initially sharp boundary

between two semi-infinite columns of liquid. A family of numerical solutions is obtained for various degrees of concentration-dependence, and some curious common-point properties of the family are noted. Methods of determining the mean diffusion coefficient and its concentration-dependence are suggested.—J. C. B.

Theoretical Problems Connected with the Determination of Autodiffusion Coefficients in Solids by the Heterogeneous Isotopic Exchange Method. G. Berthier. (*J. Chim. Phys.*, 1952, **49**, Oct., 527–536). The use of the method demands a knowledge of equations for the kinetics of the phenomena employed. The mathematical derivations are given and the results are presented in the form of tables which allow the convenient evaluation of autodiffusion coefficients. Their application is explained.—T. E. D.

The Atomistic Theory of Metallic Surfaces. C. Herring. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct., 13–19, 1951, 1–19). In view of the newness of studying surface contamination effects and adsorption, the author confines himself to a qualitative review of work on work-function and surface energy. He considers the electronic double-layer at the surface, and the resulting variation of potential energy. It is suggested that surface irregularities tend to lower the surface energy, and brief mention is made of free-electron theories of surface energy, and of the relation of the work-function to electron emission.—E. T. L.

Mechanical Effects of Interfaces. B. Chalmers. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 299–311). After summarizing mechanical properties, the effects of external surfaces, and data on crystal boundaries and interphase boundaries, the author concludes that there is insufficient knowledge to explain in detail the differences between the properties of single crystals and polycrystalline aggregates. Some effects arise in polycrystalline samples which cannot be predicted from studies on single crystals. (33 references)

Theory of Internal Boundaries. H. Brooks. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 20–64). Intercrystalline boundaries are divided into grain boundaries, non-coherent twin boundaries, coherent twin boundaries, and mosaic or small-angle boundaries. A theory is put forward which gives the energy of crystal surfaces, in terms of the dislocation theory, and on the assumption that boundaries of the first class above represent metastable states of the crystal. The elastic part of the boundary energy is computed and found to be in good agreement with experiment. (40 references).—E. T. L.

Grain Shapes and Other Metallurgical Applications of Topology. C. S. Smith. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 65–113). In two-dimensional spaces filled with polygons, the relationship $P - E + C = 1$ holds, where P, E, and C are respectively the number of polygons, edges, and corners (*i.e.*, the number of two-, one-, and zero-dimensional cells). This relationship is extended to metallography, and related to surface tension in explaining the shape of cracks on glaze, and the growth of some grains at the expense of others. (26 references).—E. T. L.

Measurement of Solid:Gas and Solid:Liquid Interfacial Energies. H. Udin. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 114–152). Methods of measurement are described, and data given on the surface tensions of copper, silver and gold. It is pointed out that the vapour phase may greatly affect the solid surface tension. The mutual wetting of metals is discussed. (26 references).—E. T. L.

Measurement of Solid:Solid Interfacial Energies. J. B. Hess. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 134–152). The author discusses particularly the deduction of relative interfacial energies from the dihedral angles between intersecting interfaces in equilibrium. Only low-energy-type boundaries consist of straight-line segments, showing that, in general, there is little dependence of interfacial energies on orientation. Miscellaneous methods of energy determination include the study of the solubility of small particles, homogeneous nucleation, the shrinking of fine wires, the creation of internal boundaries, the heat of solution of small particles, and of the energy released in grain growth. (37 references).—E. T. L.

Energies and Structure of Grain Boundaries. K. T. Aust and B. Chalmers. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct., 13–19, 1951, 153–178). This review of 85 papers deals with the amorphous cement theory, the transitional lattice theory, and the dislocation theory. The experimental evidence that is considered includes the melting of grain boundaries, the mechanical properties of the boundary, and the chemical effects there, particularly in corrosion. Experiments on tin-lead boundaries, and silicon-ferrite/silver boundaries are discussed.—E. T. L.

Boundary Migration During Grain Growth. R. L. Fullman. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 179–207). A simple theory of grain growth is outlined leading to an equation $dD/dt = As/D$ where D is the average grain diameter, s the interfacial free energy of the grain boundaries, t the time of annealing, and A is a constant at any particular temperature. On the basis of this, it is shown that the collapse of soap bubbles can be satisfactorily predicted. The author goes on to discuss the effects of temperature, inclusions, changes of boundary free-energy, adsorption, orientation, polygonization and strain on grain growth. (55 references).—E. T. L.

Interface Migration in Recrystallization. P. A. Beck. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 208–247). Thirty illustrations accompany the text which includes a discussion of the difference between grain growth and recrystallization. In the former, the boundaries migrate away from the centres of curvature, and in the latter, towards it. The author discusses the effect of strain, temperature, orientation and impurities, and considers strain-induced migration, and sub-boundary migration, including subgrain growth, polygonization and the driving energy in recrystallization. (67 references).—E. T. L.

Metal Surface Phenomena. H. H. Uhlig. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 312–335). Pointing out that a platinum wire may expand 0·0008% in length on being wetted, the author considers the process and effects of adsorption on metals, including the adsorption of oxygen on tungsten.—E. T. L.

Stress Relaxation Across Interfaces. A. S. Nowick. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 248–268). The author discusses stress relaxation at constant stress, creep at constant stress, creep recovery, internal friction and dynamic modulus. He also considers the effect of impurities and the activation energy. It is proposed that present experimental data may be explained by a theory that relaxation occurs by the transfer of atoms from one edge dislocation to another. Thus, the rate-determining process may be the creation of a vacancy at the centre of an edge dislocation. (28 references).—E. T. L.

Phase Transformations at Interfaces. A. H. Geisler. (*Amer. Soc. Met. Seminar on Metal Interfaces*, Oct. 13–19, 1951, 269–298). Many transformations are nucleated at preferential sites in the microstructure. Local formation of particles at grain boundaries of the parent phase may cause susceptibility to corrosion and brittle fracture, and hardening. The activation energy for the initial hardening is of the order of one-third of that for volume diffusion. The author also discusses a process in which nodules are nucleated almost exclusively at grain boundaries. (41 references).—E. T. L.

Application of X-Rays to Metallurgy. G. N. Ramachandran. (*Metal Market Rev.*, 1952, **5**, Sept., 7–11). A number of applications of the X-ray diffraction technique to metallurgical problems are reviewed. Among the applications discussed are the preparation of phase diagrams of alloy systems, the relationship between crystal structure and physical properties, the measurement of grain size and the measurement of stresses.—B. G. B.

Production of Monochromatic X-Radiation for Microradiography by Excitation of Fluorescent Characteristic Radiation. T. H. Rogers. (*J. Appl. Phys.*, 1952, **23**, Aug., 881–887). Materials from titanium to molybdenum have been compared, using the new beryllium-window X-ray tubes. A new degree of monochromaticity is found.—E. T. L.

A Point Focussing X-Ray Monochromator for the Study of Low Angle Diffraction. L. Shenfil, W. E. Danielson, and J. W. M. DuMond. (*J. Appl. Phys.*, 1952, **23**, Aug., 854–859). An anastigmatic point focus is obtained by means of two cylindrically-bent quartz crystals whoe focal circles are mutually perpendicular.—E. T. L.

X-Ray Diffraction by Face-Centered Cubic Crystals with Deformation Faults. M. S Paterson. (*J. Appl. Phys.*, 1952, **23**, Aug., 805–811). Calculations show that the intensity of faulting can be calculated from the breadths of reciprocal lattice streaks, or from the displacement of the intensity peaks.—E. T. L.

Evidence of a Reaction of the *n*, α Type in Iron. J. P. Longchamp. (*J. Phys. Radium*, 1952, **13**, June, 333–338). Photographic emulsions partly covered with a thin layer of iron were exposed to an intense neutron radiation. There was an excess of traces in the iron-covered part, suggesting new values for the effective cross-section of the iron and chromium atoms.—E. T. L.

Thermal Diffusion of X-Rays by α-Iron Monocrystals and Dynamics of the Cubic Centred Lattice. H. Curien. (*Acta Crystallographica*, 1952, **5**, May, 393). [In French]. A note on the propagation of waves within the monocrystal of α-iron includes curves of dispersion for waves propagated in the direction of the quaternary, ternary, and binary axes. Calculation of the dispersion curves as a function of atomic coefficients is mentioned.—T. E. D.

Aligning Single Crystals for X-Ray Diffraction. L. S. Birks and A. B. Wing. (*Rev. Sci. Instruments*, 1952, **23**, Aug., 442–443). The authors describe an aligning attachment, whereby single crystals can be optically aligned without the need for a preliminary photograph. It is attached to the stage of a polarizing microscope, so that its horizontal axis intersects the microscope axis. The crystal can be set in the field of view by a translational motion, and the goniometer head can be turned about its own axis. The attachment may be used for anisotropic crystals, including opaque crystals.—L. H.

A Simple Method of Determining Orientation in Recrystallized Wires of the Cubic System. T. S. Hutchison. (*J. Sci. Instruments*, 1952, **29**, Aug., 257–258). An optical method, using standard laboratory apparatus, is described, for estimating the type and degree of orientation in wires of the cubic system, re-crystallized from the hard-drawn condition, with a minimum grain size of 10^{-2} cm. An example of its use is given.—L. H.

The Iron-Nitrogen System. The Crystal Structure of ε-Phase Iron Nitrides. K. H. Jack. (*Acta Crystallographica*, 1952, **5**, July, 404–411). [In English]. Additional super-lattice reflexions have been observed at intermediate nitrogen concentrations and at concentrations less than Fe_3N. Proposed structures are correlated with observed unit cell dimensional changes. A mechanism for the ϵ–$F_2N \rightarrow \zeta$–Fe_2N transition is suggested. A discussion of the structural features of the iron-nitrogen system is presented.—T. E. D.

The Absorption Factor for Rod-Mounted Specimens in the Powder Method of X-ray Analysis. E. Møller and E. Jensen. (*Acta Crystallographica*, 1952, **5**, May, 345–348). [In English]. A method of calculating absorption factors is described, and tables are given of these factors as functions of θ and μr for different rod materials and values of (core radius)/(specimen radius). The influence of the variables on the intensity ratio between two X-ray diffraction lines is discussed. Rod-mounted specimens are said to be as good as homogeneous specimens for most analytical purposes.—T. E. D.

The X-Ray *K* Absorption Edges of Covalently Bonded Cr, Mn, Fe and Ni. G. Mitchell and W. W. Beeman. (*J. Chem. Phys.*, 1952, **20**, Aug., 1298–1301). An empirical correlation is derived between low-energy *K* edge absorption and empty *4p* orbital in 17 covalent complexes of chromium, manganese, iron and nickel.—T. E. D.

Diffuse Reflection of Neutrons from a Single Crystal. R. D. Lowde. (*Proc. Phys.* Soc., 1952, **65A**, 857–858). Diffuse streaks have been found in the Laue diffraction pattern of an iron single crystal placed in a beam of neutrons from the Harwell pile. These are caused by accelerated and decelerated neutrons, and the observations are in agreement with the theory of magnetic inelastic scattering.—K. E. J.

Neutron Scattering and Polarization by Ferromagnetic Materials. C. G. Shull, E. O. Wollan, and W. C. Koehler. (*Phys. Rev.*, 1951, **84**, Dec. 1, 912–921). Neutron diffraction studies are reported for a series of magnetized and unmagnetized ferromagnetic materials (including iron, cobalt, and oxides). Diffraction patterns for unmagnetized polycrystalline samples possess both nuclear and magnetic components. Studies of Fe_3O_4 support a proposed ferromagnetic structure. The formulation of the interaction function between the neutron's magnetic moment and the internal fields in a ferromagnet is substantiated. The method of monochromatic beam polarization adopted is found to compare very favourably with others in certain respects. (14 references).—K. E. J.

Investigations on Heat-Treatable Hot-working Tool Steels within the Intermediate Transformation Range. W. Eilender, R. Mintrop, and W. Lutz. (*Stahl u. Eisen*, 1952, **72**, Sept. 11, 1149–1156). After tests on the properties obtainable by heat-treating alloy steels in the intermediate transformation range, it was concluded that this procedure should be suitable for hot-working steels. Important advantages should be reduced danger of distortion and cracking, increased impact strength, greater fatigue strength at temperatures of about 500° C., and good abrasion resistance. The isothermal transformation diagrams of five alloyed tool steels have been derived ; these indicate the most suitable temperature for the intermediate transformation. The values of hardness, impact strength and temper brittleness of the various samples are compared. In certain circumstances, the properties obtained by intermediate range heat-treatment are superior to those by conventional procedures, but if the temperature of the tool exceeds certain critical temperatures, temper brittleness ensues. The intermediate structure tends to cause a more rapid onset of hot brittleness. The suitability of intermediate stage heat-treatment for tools to work at above 500° C. under high stresses needs special checking.—J. P.

Constitution Research as a Basis for the Heat-Treatment of Steels. A. Rose and W. Peter. (*Stahl u. Eisen*, 1952, **72**, Aug. 28, 1063–1075). The extension of constitution investigations to processes of hardening and heat-treating necessitates the representation of transformations in the form of time-temperature-transformation (TTT) diagrams for isothermal and steadily falling temperature conditions. The forms of the diagrams recommended give more information than the hitherto published **S**-curves. The various forms of TTT diagrams are presented for a series of steels of fixed carbon content and with alloying elements of differing affinities for carbon, namely, nickel, manganese, chromium and vanadium. All the steels exhibited pearlite, intermediate and martensite transformations, differing only in respect of the temperature ranges of transformation. With increasing carbon affinity of the alloying element, the pearlite transformation shifts to higher, and the intermediate transformation to lower, temperatures, *i.e.*, the pearlite and intermediate stage curves, which practically coincide with an unalloyed steel, become separated, and the intermediate and martensite regions approach and overlap. The distribution of the alloying element between carbides and the solid solution, and the effect of alloy addition on the carbon enrichment and decomposition of austenite are discussed.—J. P.

Carbides in Chromium, Molybdenum and Tungsten Steels. K. Kuo. (*J. Iron Steel Inst.*, 1953, **173**, Apr., 363–375). [This issue].

Transformation Characteristics of Nickel Steels. (Mond Nickel Co. Ltd. publication, 1952, 5–73). This book is subdivided into two sections. The first deals with the transformation characteristics from a structural standpoint, whilst the second gives detailed diagrams and figures on transformation and hardenability data for 13 types of nickel steels.

The Measurement of Heats of Transformation by Differential Thermal Analysis. G. Sabatier. (*Compt. Rend.*, 1952, **235**, Sept. 15, 574–575). The sample is diluted with the reference substance so that the conductivity of the mixture is sensibly that of the latter. The heat of transformation is proportional to the area under the peak of the curve, and, by calibrating the apparatus with substances of known properties, an accuracy of 5% is possible for a sample weight of 100 mg.

Influence of Sulphur on the Properties of Cementite. J. Drain and A. Michel. (*Rev. Mét.*, 1952, **49**, Aug., 585–589). The effect of sulphur on specially prepared pure cementite has been investigated by thermomagnetic analysis. Sulphur substitutes for carbon modify the Curie point, stabilize cementite, and diminish the speed of the austenite transformation. It can be removed by a desulphurizer such as magnesium.—A. G.

Metal-Metalloid Bonding in Steel and Alloys Similar to Steel. G. Hägg. (*Värm. Bergsmann Ann.*, 1951, 17–31). [In Swedish]. This is a review of published work on the metal-metalloid bond in the phases of transitional metals. Results obtained by Kuo and Hultgren regarding the distribution of alloying elements between ferrite and cementite can be explained by the tendency of the elements to bond with carbon. Alloy distribution ratio (cementite/ferrite content) should decrease as the atomic number increases. Kiessling's treatment of chromium boride with a flow of dry ammonia is discussed, the various phases observed being tabled. In each of the systems Mn-B, Fe-B, and Co-B are found the phases Me_2B and MeB. Their composition has

been determined by X-rays during equilibrium at 1100° C. giving a number of tie-lines in the two-phase range.—G. G. K.

Phase Changes Associated with Sigma Formation in 18–8–3–1 Chromium–Nickel–Molybdenum–Titanium Steel. K. W. J. Bowen and T. P. Hoar. (*Proc. First World Metallurgical Congress*, Sept., 1951, 695–706 : Amer. Soc. Met., 1952).

Graphical Study of Complex Phenomena. J.-J. Comhaire. (*Rev. Mét.*, 1952, **49**, Aug., 561–566). A geometrical method for the graphical representation of systems with *n* components is described and illustrated by the systems Mn–Fe–Cu–Si in the presence of aluminium and Fe–Mg–Cu–Si–Mn with aluminium. Suggested applications include : structural diagrams, the study of the reducibility of mixtures of impure oxides, and the viscosities of slags.—A. G.

Aims and Means of Ferrous Metallurgical Research. F. Wever. (*Stahl u. Eisen*, 1952, **72**, Aug. 28, 1053–1055). The aims of the Max-Planck-Institut für Eisenforschung in pursuing fundamental research are discussed and exemplified by reference to present work on slag-metal equilibria and isothermal transformation kinetics.—J. P.

CORROSION

The Corrosion of Metals. M. Dorizzi. (*Peintures, Pigments, Vernis*, 1952, **28**, June, 377–382). Various corrosion mechanisms are surveyed, including chemical, electrochemical, and biological. Three types of protection are then discussed ; these are surface treatment, the use of chemical inhibitors, and protective coatings.—T. E. D.

Theory of Overpotential and Mechanism of Corrosion. R. Audubert. (*J. Chim. Phys.*, 1952, **49**, July–Aug., c97–c104). General principles of the modern theory of overpotential are outlined and their application to corrosion reactions are evaluated mathematically. A number of practical results are compared with the theory.—T. E. D.

Corrosion-Resistant Materials in Marine Engineering. L. W. Johnson and E. J. Bradbury. (Inst. Marine Engineers: *Mond Nickel Co., Ltd.*, 1952). A short account of the theoretical aspect of corrosion is followed by a review of corrosive media encountered in marine engineering. This is in four parts : Fuels and products of their combustion ; boiler water and steam ; sea water ; and the atmosphere. The corrosion and properties of a very wide range of metallic materials is next discussed in detail and it is concluded that materials are available which are resistant to practically all conditions of service. Design remains one of the principal factors in overcoming corrosion and erosion problems.—B. G. B.

Corrosion—Design It Out ! R. B. Mears. (*Canad. Metals*, 1952, **15**, May 24–27). Causes of corrosion are reviewed and recommendations are made to the designers of plant with a view to reducing or eliminating it.—R. A. R.

The Cathodic Corrosion of Iron and the Anodic Formation of Crystalline Ferrite by Electrolysis of Molten Soda. M. Dodero. (*J. Chim. Phys.*, 1952, **49**, Feb., c210–c213). The electrolysis of molten soda at 500°–700° C., using iron electrodes under certain conditions, allowed the study of cathodic corrosion and of the anodic formation of crystalline sodium ferrite. The mechanisms of the corrosion, and of the ferrite formation, are explained, and the electron-transfer involved is represented graphically.—T. E. D.

Corrosion Resistant Materials. H. T. Francis. (*Machine Design*, 1952, **24**, Oct., 155–158). The author defines the corrosion problem, and explains three principles of corrosion prevention : (1) Isolation of the metal from the corrosion medium ; (2) sacrificial protection by a metallic coating ; and (3) modification of the corrosive environment. He evaluates preventive measures.—E. C. S.

Role of Crystal Orientation in the Oxidation of Iron. E. A. Gulbransen and R. Ruka. (*J. Electrochem. Soc.*, 1952, **99**, Sept., 360–367). Electron diffraction studies have shown that ferrous oxide films produced on Puron grade iron in hydrogen-water mixtures or *in vacuo* at 650–750° C. exhibit marked epitaxis, the [010] direction of the FeO growing in the [110] direction of the α-iron when the oxide film is formed on the (100), (211), and (311) planes, Films of Fe_3O_4, grown at 300–500° C. on the same planes, were also highly orientated, orientation relationships being the same as for FeO. During the transformation $Fe_3O_4 + Fe \rightarrow 4FeO$ and the reverse reaction, the orientation relationships are maintained. Rapid oxidation in oxygen produces a randomly orientated film, though if grown on already orientated FeO, the oxide rapidly takes on the orientation of the original FeO layer. These findings are discussed with respect to the general phenomena of oxidation and corrosion.—J. P.

The Isolation and Examination of Films from Metal Surfaces—An Improved Technique. T. J. Nurse and F. Wormwell. (*J. Appl. Chem.*, 1952, **2**, Sept., 550–554). The technique used at the Chemical Research Laboratory for the removal of oxide films from metal surfaces has been improved, with the objects of obtaining larger pieces of film and of facilitating their manipulation. The improved technique, entailing reinforcement of the film by Formvar (polyvinylformal) resin, is described.—E. C. S.

Methods of High Temperature Oxidation Testing and Evaluation of Observations. C. Wagner. (*Amer. Soc. Met. Special Publication*, "*High Temperature Properties of Metals.*" 1951, 93–132). Methods of determining the rate of oxidation, the structure of the oxide layers, rate laws, and evidence on the mechanism of oxidation are discussed. The effects of the concentration of the oxidising agent, temperature, and the composition are dealt with, alloys being especially studied. (89 references).—E. T. L.

Oxidation of Alloys by the Wire Life Test Method. A. de S. Brasunas and H. H. Uhlig. (*Amer. Soc. Test. Mat. Bull.*, 1952, May, 71–75). The A.S.T.M. wire life test for evaluating electric heating wire makes use of an 0·025-in. dia. wire, heated electrically for cycles of 2 min. on and 2 min. off. The test is continued until either the wire fails or the electric resistance increases by 10%. The experimental procedure for the test, which was used for rapid evaluation of alloys and various metal surface treatments for high temperature oxidation, is described. Results are given showing the effects of humidity, temperature, surface contamination, and wire pretreatment on the wire life.—B. G. B.

The Role of Oxygen in Corrosion and Cathodic Protection. R. V. Comeaux. (*Corrosion*, 1952, **8**, Sept., 305–309). The rôle of oxygen in electrochemical corrosion and its effect on cathodic protection requirements is discussed. Complete protection is achieved when electrons are supplied at the steel surface at the same rate as oxygen molecules diffuse to the surface. The application of cathodic protection to underground lines and tank bottoms at a Texas oil refinery is described.—G. P. A.

Potential Measurements in Cathodic Protection Designs. R. P. Howell. (*Corrosion*, 1952, **8**, Sept., 300–304). Criteria for determining effective cathodic protection include current density, potential change, current-potential curves and copper sulphate potential. Some case histories are discussed.

Corrosion Control in Gas-Lift Wells. D. A. Shock and J. D. Sudbury. (*Corrosion*, 1952, **8**, Sept., 296–299). Investigation of corrosion in gas-lift oil wells in Oklahoma has shown corrosion rates below the bottom gas-lift value to be as much as ten times as great as at the surface of the well. It was also found that a considerable water column was present in the annular space. An inhibitor that will mix in the water and diffuse to the bottom is therefore required.—G. P. A.

Prevention of Machinery Corrosion. (*Indust. Finishing*, 1952, **5**, Sept., 102–115). A general article describes the mechanism of corrosion and means available to combat it, as applicable to general machinery not made from corrosion-resistant alloys. Electrolytic corrosion of ferrous materials and crevice effects are particularly described. Protection methods using metal coatings, oxide and other conversion coatings, paints and temporary protectives are detailed. Illustrated examples are given of the successful use of the " Ensis " series of fluid protectives (Shell Petroleum Co. Ltd.).

Cathodic Protection at Steam-Electric Generating Stations. J. H. Collins and E. H. Thalmann. (*Corrosion*, 1952, **8**, Sept. 314–320).

Cathodic Protection against Cavitation Corrosion. V. Foltyn. (*Strojírenství*, 1952, **2**, 9, 402–408). [In Czech]. The problem of cavitation corrosion is surveyed, together with the factors facilitating its occurrence. Experiments on cavitation corrosion, employing a magnetostriction oscillator as the source of vibrations, are described, and it is shown that cathodic protection can reduce the loss in weight with time by a factor of two in most of the metals tested, including iron, the oscillation frequency being 8600 cycles/sec. Reference is made to the work of Nowotny on the effect of temperature differences between the liquid and the metal on the corrosion rate, and to the work of Petracci, who showed that, by the choice of suitable frequencies, cathodic protection can completely suppress cavitation corrosion in ordinary steel.

The Effect of Shot Peening on Damage Caused by Cavitation. N. Grossman. (*Amer. Soc. Test. Mat. Bull.*, 1952, July, 61–66). Cavitation in a liquid may result from rapid relative motion between the metal and liquid. Slight increase in pressure will collapse the vapour cavity or bubble, causing a concentrated liquid impact which may result in erosion and pitting of the metal surfaces. In the investigation described two steels and one brass were tested. The apparatus used was of the vibratory type, and details of the testing technique are given. The amount of cavitation damage, as measured by the rate of loss of weight per unit time, was decreased by shot peening. The amount of decrease varied from 7 to 58% for the different materials tested. The effect of shot-peening is to raise the yield point of the surface material without breaking any bonds. The greatest improvement occurs in the initially softest materials, provided that they can be work-hardened.—B. G. B.

Corrosion. M. G. Fontana. (*Indust. Eng. Chem.*, 1952, **44,** Sept., 101A–104A). The erosion aspect of corrosion is often underestimated when selecting materials. Erosion-corrosion test apparatus which gives results for metals and alloys, correlating well with service conditions, is described.

Corrosion. M. G. Fontana. (*Indust. Eng. Chem.*, 1952, **44,** Oct., 101A–104A). Details are given of the various alloys available for handling nitric acid under general or particular conditions.—K. E. J.

A Generalized View of the Scaling Behaviour of Metals. J. T. Waber. (*Metal Progress*, 1952, **62,** Sept., 76–80). A review of the rate of scaling of metals during oxidation at elevated temperature is made. The four proposed laws of growth, parabolic, linear, logarithmic, and cubic, are discussed. Metals which oxidize by a linear law are least desirable for high temperature use and those obeying the logarithmic or cubic laws are the most suitable. A number of deviations from the Pilling-Bedworth rule, where the ratio between the molecular volume of the oxide and the volume of an equivalent number of metal atoms is said to determine which law will be obeyed, are discussed.—B. G. B.

Atmospheric Corrosion of Low Alloy Steels. H. R. Copson. (*Amer. Soc. Test Mat. Preprint No.* 70, 1952). The results of atmospheric corrosion tests carried out on 71 alloy steels exposed for nine years in industrial and marine atmospheres are reported. The surface level was lowered only 1 mm. or less in this period. Most of the attack occurred in pits. The depth of pitting and the overall loss in weight of the sample were not complementary. Alloying decreased both pit depth and loss in weight, but the efficiency of alloying elements was more pronounced in reducing the latter. The new data confirmed the beneficial effects of fractional percentages of copper and small amounts of phosphorous. Addition of nickel produced a low initial loss in weight, and for long exposures, was particularly effective in decreasing the depth of pitting. Chromium was found particularly effective in flattening the weight-loss/time curve. Chromium and copper were more effective in the industrial atmosphere, and manganese was more effective in the marine atmosphere.—B. G. B.

The Caustic Embrittlement of Steel. J. Vodsedálek. (*Strojírenství*, 1952, **2,** 9, 410). [In Czech]. The causes and prevention of corrosion of carbon and alloy steels and non-ferrous alloys in alkaline solutions are considered, and modern theories are reviewed. Data relating to the influence of stresses, particularly near welds, on the influence of currents, mainly for anodic protection, and on the effect of temperature and concentration in the case of attack by sodium hydroxide solutions, are given. Embrittlement can be avoided by : (1) reducing internal stresses below the limit of corrosion sensitivity ; (2) increasing this limit by rolling and spherodizing the surface ; (3) anodic protection ; (4) the use of aluminium or chromium, or nickel and chromium as alloying elements ; and (5) reducing the activity of the corroding electrolyte by adding inhibitors and using Schröder's detectors for activity control.—P. F.

Rupture of Ferrous Metals by Corrosion and Associated Mechanical Stresses. E. Herzog. (*Métaux–Corrosion–Indust.* 1952, **27,** Sept., 329–357). The mechanism of caustic cracking is discussed and the effect of nitrate solutions and of temperature on the rate of failure of test samples is reviewed. Investigations into the effect of variation in composition, structure, and treatment (*e.g.*, rolling) of ordinary mild steels on their resistance to stress corrosion cracking are described. This type of stress corrosion cracking also occurs in austenitic steels with less than 13% Cr, and several examples of such failure are given and discussed. The combined effect of corrosion and fatigue is also considered. (25 references).

Intergranular Corrosion in Austenitic Stainless Steels Resistant to Acids, and Methods of Determining their Susceptibility to Corrosion. B. Hermelin. (*Acciaio Inossid.*, 1952, **19,** Sept.–Oct., 99–108). [In Italian]. After reviewing typical austenitic steels, the author discusses the improved resistance to corrosion obtained by additions of from 1 to 6% of molybdenum. The causes of intergranular corrosion are examined and methods of diminishing it are considered.

Corrosion Resistance of Sprayed-on Austenitic Materials. O. van Rossum. (*Metalloberfläche*, 1951, **5,** Aug., A113–A115). Chemical analyses, hardness tests, examination of micrographs, corrosion tests in boiling 40% HNO_3 and glacial acetic acid, and tests on grain disintegration in acid copper sulphate solution were carried out on sprayed surface materials, primarily austenitic steels of the Cr–Ni–Mo–Ta–Nb and Cr–Ni–Mo–Ti types. It is shown that, apart from the porosity of the layers, a good corrosion resistance can only be achieved when the possibility of oxidation can be largely eliminated. The austenitic deposits were found to be liable to attack by acids, which was facilitated by oxides in the coating formed during the spraying process.—P. F.

On the Passivity of Stainless Steels in Acid Media. J. M. Defranoux. (*Rév. Mét.*, 1952, **49,** Sept., 664–672). The establishment and stability of passivity under ranging conditions of initial surface, and oxidizing power, are investigated. Passivity is only permanent when the oxidizing power exceeds a certain limiting value. In highly acid media, a special type of corrosion occurs accompanied by a very high potential and termed the transpassive state. In this case corrosion rate is constant and independent of initial surface state.—A. G.

Protection of Metallic Surfaces Against Corrosion. P. Tyvaert. (*Inst. Supérieur des Matériaux et de la Construction Mécanique*, 1951, 1–86). A detailed review of methods, other than electrolytic, of protecting metallic surfaces is presented. The preparation of the surface by various methods of degreasing and descaling are described in some detail. The use of enamel coatings is discussed. The application of coatings of zinc, tin, lead, or aluminium by immersion of the object in a bath of the molten metal, or by spraying the object with the molten metal from a gun is also discussed. A number of other types of protective treatment, including Sheradizing, are described.—B. G. B.

Vapour Phase Inhibitors. H. L. Bennister. (*Research*, 1952, **5,** Sept., 424–432). The chemical composition and mechanism of action of vapour phase corrosion inhibitors are described. Their outstanding feature is that they can exert their influence on the surface to be protected, over a distance. They may be used in powder form, solution, or coated on paper. The chief applications are for aircraft and other engines having inaccessible surfaces, and for anti-corrosive storage and packaging of a large variety of mechanisms and equipment. (17 references).—K. E. J.

Corrosion by Liquid Zinc of Cast Irons with and without Previous Oxidation. P. Bastien and P. Azou. (*Fonderie*, 1952, Mar., 2831–2845). Of the cast irons tested, grey iron was attacked least by molten zinc. Malleable cast iron also behaved well, whereas mild steel was very heavily attacked. An addition of 2·5% Mo seems to prevent attack by zinc. An oxide layer on the iron will prevent attack by zinc if the layer has high resistance to thermal and mechanical shock, is impermeable to metal ions, and has good adherence to the iron. The formation of intermediate layers of $FeZn_{13}$ and $FeZn_7$ is discussed.—R. A. R.

The Resistance of Cast Irons to Heat and Corrosion. G. Somigli. (*Fonderia Ital.*, 1952, **1,** Sept., 122–136). The author studies the behaviour of cast irons when submitted to (*a*) heat and (*b*) the action of corrosive liquids and gases. Brief reference is made to special irons. (66 references).

Corrosion Testing—A Tool of Conservation. E. G. Holmberg. (*Chem. Eng. Prog.*, 1952, **48,** Aug., 377–380). A method of testing specimens within vessels or pipelines is described. The behaviour of various alloys under various test conditions are mentioned, and the appearance of $\frac{1}{2}$ in. dia. × $\frac{1}{16}$ in. thick specimens, after exposure to sulphuric acid for 45 days at 170–190° F. and 190–210° F., is shown.—T. E. D.

New Standard for Coating Thickness. M. Kronstein. (*Chem. Eng.*, 1952, **59,** June, 201–203). After measuring the profile of grit-blasted steel surfaces and determining the effect

of various thicknesses of organic coatings on moisture penetration, it was concluded that coating thicknesses should be three times the profile depth to resist moisture penetration. The electrographic printing method was used to detect moisture penetration and under-film spreading of moisture.

ANALYSIS

Analytical Chemistry as a Metallurgical Tool. W. Koch. (*Stahl u. Eisen*, 1952, **72**, Aug. 28, 1056–1063). Progress in analytical chemistry has developed more sensitive routine test methods which now permit determination of many trace elements. Spectrographic methods are now very important and have many advantages over other procedures. The examination of oxide inclusions, isolated from steel, by analytical as well as by microscopical and X-ray methods is quoted as an example of how this progress in analytical chemistry is being profitably applied.—J. P.

Contribution to Ferrous Metallurgical Chemistry. (Second Half of 1951). P. Klinger. (*Stahl u. Eisen*, 1951, **72**, Oct. 9, 1309–1312; Oct. 23, 1369–1372). The literature on developments during this period are reviewed. Subjects covered include ion-exchange resins for purification and separation of inorganic and organic reagents, apparatus and equipment for analysis, newly developed methods for qualitative and quantitative examination of metals, ores, slags, and fuels, separation and examination of carbides, flame photometry and corrosion testing. (54 references).—J. P.

Development of Analyical Chemistry in the U.S.S.R. in 1949. Iu. A. Klyachko. (*Zavodskaya Laboratoriya*, 1950, No. 5, 515–526). [In Russian]. A comprehensive review is given, with 84 references, of analytical methods for ferrous and non-ferrous metallurgy developed recently in the U.S.S.R.—S. K.

Application of Extraction in Chemical Analysis. A. K. Babko. (*Zavodskaya Laboratoriya*, 1950, No. 5, 527–536). [In Russian]. A study is made of the process of extraction with organic solvents in relation to analytical methods used in ferrous and non-ferrous metallurgy. Extraction is shown to have several advantages over precipitation in these fields.—S.K.

Apparatus for Moisture Determination by Dielectric Permeability. L. A. Rakhmanov. (*Zavodskaya Laboratoriya*, 1950, No. 5, 582–584). [In Russian]. A special condenser and circuit are described for the determination of moisture in powdered or granular industrial materials to an accuracy of 0·2% over a range of 4–40% by the dielectric method.—S. K.

Developments in the "Dead-Stop End Point" Titration Technique. J. T. Stock. (*Metallurgia*, 1952, **46**, Oct., 209–212). The theoretical aspects of the 'dead-stop end point' technique are considered, and it is concluded that the end point depends upon electrolytic oxidation at the anode coupled with electrolytic reduction at the cathode. Various new titrations for the determination of metals are discussed.

A Recording Polarograph. H. Hohn. (*Metall u. Erz*, 1943, **40**, July, No. 13–14, 197–204). After outlining briefly early experiments, the author describes the construction and applications of a direct reading polarograph, using a recording pen, which is claimed to be accurate, and is robust and easy to operate.—L. M.

Some Applications of Photomultiplier Tubes to Spectrographic Analysis. J. K. Brody. (*J. Opt. Soc. Amer.*, 1952, **42**, June, 408–415). A two-tube photomultiplier tube mounting is first described, built for a Baird 3-meter grating spectrograph, which is fitted into the plateholder position, and can be changed from plates to direct-reading in 1 min. Each tube holder can be rotated, and each tube moved with respect to its slit. A range of scanning speeds of 0·2 to 16 mm./min. is available. Amplifying is by a tuned 60-cycle A.C. preamplifier, rectified by a 60-cycle vibrator. The recorder may be used to determine the ratio of two currents, or as a microammeter. Location of lines is by scanning, after which the position of the tube is noted for re-location. Three examples of its use are given.—L. H.

An Automatic Thermal Analyses Recorder. R. L. Cunningham, H. M. Weld, and W. P. Campbell. (*J. Sci. Instruments*, 1952, **29**, Aug., 252–253). An inverse-rate recorder is described, which is compact and inexpensive, as a standard commercial potentiometer is used. The record is a series of parallel straight lines, the length of a line being proportional to the time taken to pass through a specific e.m.f. interval, so that any sudden change in line length corresponds to a transformation. Constructional details and circuit diagrams are given.—L. H.

Methods of Preliminary Chemical Analysis for Carbon, Chromium, Nickel and Molybdenum. R. H. Jacoby. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 260–265). Descriptions are given of the rapid chemical methods of determining carbon and alloying elements during the melting cycle, which are routine practice at the Key Co., East St. Louis. Nickel, molybdenum, and chromium (up to 5%) are determined by the photelometer, and carbon and chromium (more than 5%) volumetrically.—G. F.

Preliminary Analyses for Carbon, Phosphorus, and Sulphur in Steel. L. H. Arner and H. H. Johnson. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 266–271). The authors describe the rapid analysis methods for carbon, sulphur, and phosphorus in steel, which are used at the Sharon works of the National Malleable and Steel Castings Co. Carbon and sulphur are determined by combustion, and phosphorus alkalimetrically.—G. F.

Analytical and Wage Standardization in the Chemical Laboratory. C. M. Nettles. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 271–280). A description is given of the standardization of analytical procedures and the introduction of a wage incentive plan in the chemical laboratory at the Houston works of the Texas Electric Steel Casting Co. Details of the analytical methods and of the incentive system are given in appendices.—G. F.

The Determination of Arsenic in Ferro-Tungsten. H. Wirtz. (*Metall u. Erz*, 1943, **40**, Feb., No. 4, 67–68). The author criticizes the use of hydrofluoric acid in the preliminary solution of ferrotungsten, as varying amounts of arsenic are thereby volatilized.—L. H.

Accurate and Simple Determination of the Nickel plus Cobalt Content of Ferronickel. K. Wagenmann. (*Metall u. Erz*, 1943, **40**, Apr., No. 7–8, 133–134). Ferronickel normally contains cobalt to the extent of 0·1–2·0% of the nickel present. When this is added as an alloying addition to steel, the small amount of cobalt in the steel behaves practically as if it were nickel. For this purpose then, the author suggests that the ferro-nickel can be evaluated by stating the sum of the nickel + cobalt contents. He describes an accurate method of estimation, not involving the preliminary separation of iron. The nickel (contaminated with cobalt) is separated with dimethylglyoxine and the remaining cobalt with α-nitroso-β-naphthol, and the estimation is completed electrolytically.—L. H.

Determination of Oxygen in Metals and Metal Oxides by the Isotopic Method. A. D. Kirshenbaum and A. V. Grosse. (*Amer. Soc. Met.*, 1952, *Preprint No.* 39). Since the ordinary methods are tedious, the author emphasizes the speed of this method, involving the use of O^{18}. A known weight of substance to be analysed is heated with a known weight of a metal (or its oxide) containing a known amount of O^{18}, for 30 to 60 min. at near the melting point of the metal. All the oxygen is then distributed, to give an equal O^{16}/O^{18} ratio in both metals. This ratio is then rapidly determined in a mass spectrometer.—E. T. L.

Considerations Concerning the Determination of Oxygen and Inclusions in Steels. J. Varetto and L. Lacomble. (*Rev. Univ. Mines*, 1952, Series 9, **8**, July, 270–274). The accuracies of the Gotta and vacuum-fusion methods for determining oxygen in steel have been investigated. For very small oxygen contents (as occur in manganese steels) the vacuum-fusion method is to be preferred. The authors have modified the usual vacuum furnace to reduce errors due to adsorbed gases on the walls of the vessel being desorbed on heating; nearly all the surfaces are now water-cooled.—B. G. B.

Rapid Analysis of Hydrogen in Molten Steel by Vacuum Fusion Method. Y. Ishihara and S. Sawa. (*Proc. First World Metallurgical Congress*, Sept., 1951, 247–255; Amer. Soc. Met., 1952). A rapid analytical method for hydrogen in steel is described. The method involves vacuum fusion of the steel sample and subsequent measurement of the thermal conductivity of the extracted gas.—D. L. C. P.

Separate Determination of Organic Sulphur Compounds in Gases. F. M. Rapoport. (*Zavodskaya Laboratoriya*, 1950, No. 5, 560–565). [In Russian]. On the basis of information in the literature and original experimental data, a method has been developed for the quantitative determination of carbon disulphide, thiophene and carbonyl sulphide in coke-oven, producer, and other industrial gases. The accuracies of the determinations are ±5, ±5 and ±1%, respectively.—S. K.

Spectro-Chemical Determination of Chromium in Carbide Residues. I. M. Veselovskaya and V. G. Koritskii. (*Zavodskaya Laboratoriya*, 1950, No. 5, 632–633). [In Russian]. A spectrum method for the determination of chromium in the carbide residues of chromium steels is described. Standards, prepared by electrochemical solution, were chemically shown to contain 8·27, 18·66 and 34·27% of chromium, and were used for deducing the working curve. The residues were briquetted under a pressure of 3·5 tons/sq. cm., using 300-mesh copper as binder, 30 mg. of residue being used per briquette. A briquette 6 mm. long and 7 mm. in diameter was exposed to a condensed spark produced by Raiskii's scheme with a carbon auxiliary electrode. The line pair Cr 2830·47 and Cu 2824·37 was used, giving a mean relative error of 2·5%. Chromium determinations in 77 specimens enabled the phase diagram of the system Fe–Cr–C to be substantially refined.—S. K.

Spectrographic Measurement of the Diffusion Constant of Carbon in a Case-Hardened Steel. A. Camuñas Puig, and E. Asensi Alvarez-Arenas. (*Inst. Hierro Acero*, 1952, **5**, July, 768–785). [In Spanish].

Standard Methods of Coal Analysis. J. G. King. (*Coke Gas*, 1952, **14**, Oct., 363–367). The measure of agreement that has been reached by the Technical Committee 27 of the International Standards Organization (ISO/TC27) on standardization of coal-testing methods is surveyed.—T. E. D.

Spectrochemical Determination of Fluorine in Slags. J. Gillis, J. Eeckhout and N. Kemp. (*Rev. Univ. Mines*, 1952, Series 9, **8**, July, 284–288). The measurement of the intensity of a line of the CaF_2 spectra (5298·6 Å) forms the basis of the method which is described. A continuous arc of 9 amp. is used for excitation. The quadratic error of the determination is 8–10% for fluorine contents above 1% and 15–20% for contents below 1%. Fluorine contents down to 0·05% can be determined.–B. G. B.

Analytical Methods for the Determination of Cyanates in Plating Wastes and Effluents from Treatment Processes. B. F. Dodge and W. Zabban. (*Plating*, 1952, **39**, Apr., 381–384). A method is described which involves the decomposition of cyanate by acid, distillation of ammonia and estimation by Nessler reagent. The cyanate is separated from solution by precipitation with silver nitrate ; when cyanide is present, this is also precipitated, but, when treated with acid, this gives only hydrocyanic acid and no ammonia.—J. P.

Nomogram for Analysis of Watts Type Nickel Plating Solutions. I. Goldman. (*Metal Finishing*, 1952, **50**, Nov., 76–77). A nomogram is presented for converting the strength of a solution of $AgNO_3$, determined by titration, into grammes per litre of $NiCl_2 . 6H_2O$, similarly for NaCN into grammes per litre of total $NiSO_4 . 6H_2O$.—R. A. R.

Methods for the Analytical Control of Potassium Cyanide Plating Baths. W. Savelsberg. (*Metalloberfläche*, 1951, **5**, Aug., B116–B117). The analysis of complex potassium cyanide salts, containing a metal such as silver, zinc, or cadmium, by a rapid method is described. A small quantity of the liquid from the bath is titrated against Na_2S solution, and the metallic sulphide is precipitated. Lead acetate solution is used as end-point indicator, sensitive to any remaining divalent sulphur ions.—P. F.

HISTORICAL

Henry Clifton Sorby. A Pioneer in Metallurgical Microscopy. E. N. Simons. (*Metallurgia*, 1952, **46**, Oct., 199–200). The life and work of Sorby, with particular reference to his application of the microscope to the study of geological and metallic specimens, is reviewed.—B. G. B.

The History and Development of the House of Tata. Sir Frederick James. (*Bull. Inst. Met.*, 1952, **3**, Sept., 22–28). An historical account of the growth of the Tata group of concerns from their establishment in 1887 also describes aspects of India's economy. After success with textiles, Tata established the iron and steel works at Jamshedpur, the Bombay hydro-electric system, and the Indian Institute of Science. Present interests include cement, vegetable oils, chemicals, air-ways, engineering, and philanthropic, technical, and educational foundations.—K. E. J.

A Page from the Industrial Expansion of Liége up to the XVIIth Century. P. Harsin. (*Rev. Univ. Min.*, 1952, Series 9, **8**, Dec., 458–461). An account is given of the metallurgical activities of the Besche family in Sweden and in France.

The Woodcuts in Georgius Agricola's Book on Mining. W. Pieper. (*Metall u. Erz*, 1944, **41**, Mar., 49–56). This article is mainly a review of Dr. Georgius Agricola's XVIth century book "De Re Metallica". The value of this book as an historical document lies principally in the large number of carefully executed and detailed drawings. Many of these are here reproduced and annotated.—L. H.

ECONOMICS AND STATISTICS

Outlook for Electric-Furnace Scrap. E. L. Solomon. (*Amer. Inst. Min. Met. Eng., Prov. Elec. Furn. Steel Conf.*, 1951, **9**, 11–14). The author outlines the causes of the American scrap shortage and suggests possible means of overcoming it. Particular emphasis is given to scrap used in electric steel-making.—G. F.

Availability of Alloys and Allocation Procedure. F. F. Franklin. (*Amer. Inst. Min. Met. Eng., Proc. Elec. Furn. Steel Conf.*, 1951, **9**, 5–10). The author briefly discusses the availability of alloying elements in America and outlines the procedure for their allocation and distribution.—G. F.

Metals for Defense in the E.C.A. Countries. P. van de Rest. (*Proc. First World Metallurgical Congress*, Sept., 1951, 91–99 : Amer. Soc. Met., 1952). The production of ferrous and non-ferrous metals in western Europe is reviewed with mention of output, raw materials, modernization and expansion, and the Schuman Plan.—D. L. C. P.

Metals for Defense in the Free World. C. Williams. (*Proc. First World Metallurgical Congress*, Sept., 1951, 100–112 ; Amer. Soc. Met., 1952). The capacities of the countries of the free world to produce raw materials and metals are reviewed. Fabrication, manpower, and efficiency are touched upon, and comparison is made with production within the Soviet orbit.—D. L. C. P.

Metals Conservation and Substitution for Defense. K. P. Harten. (*Proc. First World Metallurgical Congress*, Sept., 1951, 113–123 : Amer. Soc. Met., 1952). The author considers the means of economizing in essential metals which are, or may be, in short supply, with examples from German experience.—D. L. C. P.

Mineral and Metal Industries of India. D. P. Antia. (*Proc. First World Metallurgical Congress*, Sept., 1951, 166–169 : Amer. Soc. Met., 1952). The author briefly describes India's mineral and metal production, imports, exports, her metal needs, and the factors affecting future development of her metal industries.—D. L. C. P.

The Iron Ore Industry in 1951. (*Usine Nouvelle*, 1952, Oct. 9, 23–25). French production of iron ores during 1951, including figures from the various localities, is given. Trends in production from these areas since 1929 are pointed out.

The Position of Algerian Metallurgy. (*Usine Nouvelle*, 1952, July 24, 19 ; July 31, 17–19). The Algerian ferrous and non-ferrous industries are surveyed. Difficulties in development are outlined and credit restrictions are mentioned. A development plan is projected.—T. E. D.

The Belgium Iron and Steel Industry in 1951. A. G. Lefebvre. (*United Nations Economic Commission for Europe, General, E/ECE/Steel/68*, July, 1952, 19–20). A short account of recent outputs and developments in the Belgium iron and steel industry is given.—B. G. B.

MISCELLANEOUS

Television in Metallurgical Industries. (*Foundry Trade J.*, 1952, **93**, Oct. 9, 415–416). The television applications cited include the control and observation of a conveyor belt along its length, scrap baling, and observation of the stock-line in a blast-furnace. A television camera is more sensitive than the human eye to infra-red light, and it can be constructed to withstand extreme temperatures and pressures.—E. T. L.

Principles and Methods of Telemetering. T. Wikland, R. Ferngren, and D. Hartman. (*Tekn. Tidskr.*, 1952, **82**, Sept. 2, 693–704). [In Swedish]. Applications of telemetering in industry and some system fundamentals are described. Signal transmission by means of frequency division and channel filtering is discussed in relation to problems of noise level and measurement errors due to distortion. Typical circuits are shown for power network telemetering.—G. G. K.

The Introduction of Statistical Methods to Industry. B. P. Dudding. (*Applied Statistics*, 1952, **1**, Mar., 3–20). The author deals with the introduction of statistical methods to factory production and their culmination in the control chart technique, and describes the advantages that accrue.

The Plant Engineer—His Functions and Responsibilities. H. E. Pridmore. (*Foundry*, 1952, **80,** Oct. 162, 164).

Metallurgical Education, Research and Development Work in Norway. A. B. Winterbottom. (*Proc. First World Metallurgical Congress*, Sept., 1951, 294–297 : Amer. Soc. Met., 1952). Until recently, Norway had no public or national bodies carrying out research and development work. Now the Metallurgical Institute has, in addition to educational staff, well-equipped laboratories and services available for use by industry.—D. L. C. P.

The General Organization of Technical and Scientific Societies in Great Britain and Switzerland. (*Sheet Metal Ind.*, 1952, **29,** July, 600–606). This is a report of the Joint Technical Conference held in Zürich during a visit of the Sheet and Strip Metal Users' Technical Association.—R. A. R.

The Steel Plant Laboratory, Its Function and Value. S. Bianco. (*Amer. Iron Steel Inst. Regional Tech. Meetings*, 1950, 141–149). The author describes the functions of the various departments of the steelplant laboratory, from the placing of the order to the point where the steel is put to its final use. The chief function and main value of the laboratory is to maintain close control over the entire process.—G. F.

Physics in the Service of Metallurgy: Major Developments Contributing to Success. Sir Andrew McCance. (*Iron Coal Trades Rev.*, 1952, **165,** July 25, 215–216, 209). The author describes some of the contributions of physicists which have assisted the study of metallurgy, and discusses the development of spectrographic analysis, the application of X-rays, and the fundamental properties of solids. Particular reference is given to the value of the Pt/Pt–Rh thermocouple.

Distribution of Research Effort in Britain. (*Nature*, 1952, **170**, Sept. 20, 465–468). A survey by the Industrial Research Committee of the Federation of British Industries, covering 301 firms is discussed with reference to the wise direction of research effort.—A. G.

On the Study of Technical Viscometer. S. Iwanami. (*Memoirs of Faculty of Technology, Tokyo Metropolitan University*, 1952, No. 2, 1–40). The author reports on the characteristics of viscous flow through pipes, short tubes, and orifices.—R. A. R.

British Cast Iron Research Association. (*Engineering*, 1952, **174**, July 11, 53). A description is given of the new laboratories of the British Cast Iron Research Association, Alvechurch, Birmingham. These comprise a chemical laboratory in an advanced state of completion, a spectrographic laboratory containing a quantometer, sand-testing, and sand research laboratories, and stores.—M. D. J. B.

New Laboratories at King's College, University of London. (*Engineering*, 1952, **174**, July 4, 23–24). A brief account is given of the opening ceremony of the new physics and engineering laboratories at King's College, London.—M. D. J. B.

The Selection and Training of Personnel. E. E. Moore. (Amer. Iron Steel Inst. : *Blast Furn. Steel Plant*, 1952, **40,** Oct., 1200–1204).

Water in Works. R. Colas. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 121–128). All aspects of water supplies in industry are discussed. The chief users are paper manufacturers and iron and steel works. Water resources and conservation are mentioned and the importance of planning demand to fit supply is emphasized. Water treatment and purification are discussed, and the circulation of water within a typical iron and steel works is illustrated. Relative consumption of water in the various departments is outlined together with the economics involved.—T. E. D.

Accident Prevention in Steel Works in the U.S.A. (*Machinery*, 1952, **81,** Nov. 28, 1136–1137). **Accident Prevention in U.S. Steel Works.** (*Times Review of Industry*, 1952, **6,** Nov., 46). A report on accident prevention in U.S. steelworks is summarized. The standard of machine guarding in the U.S.A. is not as high as in this country, but layout and spacing of machinery are better. Far more attention is paid to personal safety in the U.S.A., and the issue and wearing of protective clothing has reduced the accident rate considerably. Accident prevention publicity plays a large part. Several recommendations are made .—T. E. D.

The Safety Campaign in American Industry. H. S. Heidberg. (*Stahl u. Eisen*, 1952, **72,** Sept. 11, 1173–1174). The history of the development of safety precautions in American industry, and the methods used to achieve freedom from accidents, are discussed.—J. P.

The Problem of the Prevention of Works Accidents and Occupational Diseases. J. Troyon. (*Usine Nouvelle*, 1952, Special Spring Issue, May, 185–187). French legal aspects are outlined, and the responsibility of the employer in providing safeguards is indicated.—T. E. D.

Colour and Light in Industry. R. F. Wilson. (*Elect. Rev.*, 1952, **151,** Sept. 26, 652–654). An outline of general principles is given in the selection of colours and lighting schemes in industry. The effect of the type of artificial lighting on the colour is discussed.—L. H.

Lighting and Accident Prevention. M. A. McTaggart. (*Elect. Rev.*, 1952, **151,** Sept. 26, 661–663).

Light Intensity—The Primary Standard and the New International Unit. A. G. Thomson. (*Elect. Rev.*, 1952, **151,** Sept. 26, 657–660). A short historical survey of the development of light standardization is given, leading to the adoption in 1948 of the 'candela' as the unit of light, and methods of measuring light intensity are described and discussed.—L. H.

Progress in the Field of Works Heating. E. Krebs. (*Stahl u. Eisen*, 1952, **72,** Aug. 28, 1088–1093). On account of excessive heat consumption and the particular disadvantages of central steam heating, works heating by hot water has been provided. The possibilities of using waste heat from furnaces and machines are discussed, and the new central hot water plant, the piping system and circulation are described.

Laboratory Air Conditioning. A. E. Stacey, jun. (*Amer. Soc. Test. Mat. Preprint No.* 121, 1952). A number of different methods of ventilation suitable for laboratories are considered.—B. G. B.

Air Pollution Abatement in the Steel Industry. C. A. Bishop. (*Amer. Iron Steel Inst. Regional Tech. Meetings*, 1950, 245–261). The air pollution abatement requirements for the steel industry, as set out in a Smoke Control Ordinance of Allegheny County (Pa.), are discussed, and the progress made in studying the blast-furnace, open-hearth, and Bessemer converter problems is emphasized. The fundamentals underlying the collection of dust and fume particles are outlined, and various types of gas-cleaning plant are described.

Electrostatic Filtration: Dust Comes Out of the Air. (*Steel*, 1952, **130,** May 4, 105). The principles of electrostatic filtration are briefly described. For particles and bacteria under 5μ this method must be used instead of mechanical filters. Clean air should reduce some maintenance and operating costs in almost any industry.—A. M. F.

Standards for Steel Development. (*Economist*, 1952, **164,** July, 36–38). The report of the Anglo-American Council on Productivity on the American steel industry is discussed.

BOOK NOTICES

BOSWORTH, R. C. L. "*Heat Transfer Phenomena.*" The Flow of Heat in Physical Systems. 8vo, pp. xii + 211. Illustrated. Sydney, N.S.W., Australia, 1952: Associated General Publications Pty., Ltd.; New York: John Wiley and Sons, Inc. (Price 42*s.*)

The book starts with a clear exposition of the fundamental properties of the processes for the diffusion transfer of electric charge, heat, matter, and momentum, based on the concept of moving carriers. When, as is usually the case, these have a short mean free path compared with the distance of transport, Ohm's Law or its analogue is obeyed, but in other cases, as in radiant heat transfer, the transport is no longer proportional to the gradient of the 'voltage.' This reduction to first principles is very valuable in showing the causes for exceptions to the usual rules.

A simple account of kinetic theory for the thermal conductivity of gases leads to the experimental values and their application in the katharometer, which is the basis of well-known CO_2 meters. In the case of thermal radiation, the basic laws are stated without derivation, and some of the diagrams are so simple as to be hardly necessary. Criticism of the optical pyrometer (p. 46) does not take into account the latest developments with flat filaments. The basic equations for radiative geometrical factors are supplemented by a useful description of an optical model, but in the section on flame radiation the reason why soot gives flames non-grey-body properties is replaced by the word 'peculiar,' and non-luminous radiation, which is well understood, is barely mentioned.

Conduction in solids and liquids is treated after radiation, because it can involve two different mechanisms operating simultaneously. Convection is discussed in terms of the stagnant film theory and dimensional analysis, but no reason is given for the choice of the most useful set of dimensionless products. The chapter on the simultaneous transport of heat and matter (as in conduction and evaporation) shows how the greater physical complexity of this subject has hindered the adequate development of empirical dimensionless formulæ. In the next chapter the electrical analogue is developed as a practical method of solving the heat transfer equations, but the introduction of a thermal ohm seems to be taking it a little too far. The book concludes with a brief discussion of similarity laws for thermal models.

To give a good fundamental background to the subject and make the reader familiar with physical ideas, this book is highly to be recommended. It is readable and clear, and fills a very useful place beside the treatises which give a more final statement of the best empirical formulæ for each given case.—M. W. THRING.

BRITISH ELECTRICAL DEVELOPMENT ASSOCIATION. "*Lighting in Industry.*" (Electricity and Productivity Series No. 2). 8vo, pp. xii + 154. Illustrated. London [1952]: The Association. (Price 9*s*.)

"A man works best when he can see what he is doing" is the theme of "Lighting in Industry," a new book in the Electricity and Productivity Series, published by the British Electrical Development Association.

The book deals with works lighting from the user's point of view, and provides essential data for the planning of good industrial lighting. The result of an increase of illumination on productivity in various trades is recorded in Chapter 1, followed by economic considerations and a guide to the surveying of lighting installations, with suggestions for the improvement of existing conditions, in Chapter 2.

Chapter 3 deals with particular applications, including drawing offices and production areas, and gives useful information on the important and special lighting required for inspection. The use of polarized light, and the transmission of light into the interior of vessels containing dangerous vapours by means of internal reflections in glass or perspex rods, is included.

Chapter 4 deals with special conditions in various industries, including foundries and metal working shops. Some of the points discussed for other industries are, however, applicable to steelworks.

The full benefit of a good lighting installation can be obtained only by adequate maintenance, but maintenance is costly and, if overdone, can be uneconomical. Chapter 5 gives the answer to this problem by providing data for an economical maintenance schedule. The effect of switching on economy is also discussed.

Chapter 6 brings forward a point which is often overlooked, the effect of correct colour of interiors on lighting. It rightly points out that colour treatment is a major factor in the success or failure of a scheme.

The concluding chapter and the appendices include data required to design a lighting installation and calculate running costs, Appendix 2 being a reminder that adequate factory lighting is a statutory obligation.

The book forms a clear and logical guide to good industrial lighting, which should be studied by all executives responsible for works production and lighting.—G. W. LEVEY.

FUCHS, E., and H. BRADLEY. "*Welding Practice.*" Volume II. "*Welding of Ferrous Metals.*" La. 4to, pp. xii + 198. Illustrated. London, 1952: Butterworth's Publications, Ltd. (Price 22*s*. 6*d*.)

This volume is the third in order of appearance in a series of three dealing generally with the practice of welding. Volumes 1 and 3 are complementary to it, and deal with "Welding Methods and Tests" and "Welding of Non-Ferrous Metals," respectively. All three volumes originally appeared in pamphlet form for internal use in the works of the Imperial Chemical Industries, Ltd., and many of the leading technicians of that organization participated in their preparation. The result is a very clear, comprehensive, and authoritative series.

Volume 2 is divided into seven chapters, and each chapter is, as far as is possible, built up around five main points, namely, metallurgical factors, design data, welding materials, workshop practice and special applications of welding. Whilst much has already been written on these subjects, this volume serves the useful purpose of bringing together between two covers the most important points to be considered in the welding of ferrous metals; the latest codes and standards are referred to, and the quality and clarity of its profuse illustrations and tables make it an outstandingly useful book.

The welding of mild steel by the metallic arc process is dealt with in the first chapter, and, whilst design aspects receive the largest proportion of the space allotted to this subject, a useful section is devoted to classification and characteristics of arc welding electrodes. Chapter 2 follows on, naturally, with the welding of mild steel by processes other than the metallic arc. Here, each process is considered in connection with its best field of application. Undue prominence is given to water-gas welding, whilst nothing is said about argonarc welding, a process with interesting potential fields of application. Resistance welding is adequately covered, and some excellent photographs of actual applications and of sections of welds are presented.

In Chapters 3 and 4, information is given on the welding of low-alloy and chromium-nickel steels, respectively. As would be expected, the metallurgical side of the question is predominant, because successful welding depends largely upon what account is taken of welding materials, procedure and heat-treatment. The success of the low-hydrogen or basic coated type of electrode in preventing cracking when welding low-alloy steels is well written up, giving not only a clear impression of the present position, but also of the work which led up to its development. As an example of weld decay in 18/8 stainless steel, a photograph of our old friend the welded ball float is given. A great deal is said, however, on the causes and prevention of weld decay.

Cast iron welding is dealt with in Chapter 5, in which attention is rightly drawn to the fact that the best all-round method is the oxy-acetylene torch, using a cast-iron filler rod; this method should be used whenever its preheating and cooling requirements permit it.

A most impressive survey of the pros and cons of pressure-vessel welding is given in Chapter 6. The comprehensive treatment of the relevant design details is outstandingly good, and this alone should be sufficient to assure the success of the book. Chapter 7 goes on to deal with the various applications of welding to hard-surfacing.

The book is very well balanced, and this, coupled with its technical accuracy and clarity of presentation in both the text and illustrations, enables it to be commended to all those interested in welding, particularly designers and welding engineers.—A. J. HIPPERSON.

SMITH, D. M. "*Visual Lines for Spectroscopic Analysis.*" Second Edition completely revised 1952. 8vo, pp. 102. Illustrated. London, 1952: Hilger and Watts, Ltd. (Price 16*s*.)

This is a text-book for the training of craftsmen in spectroscopy. How would you like to have a colleague in your laboratory to whom you could take any odd sample—the bit of steel sent to illustrate a complaint, a precipitate found unexpectedly during analysis, the inclusions from another bit of steel, a flake of paint, a scrap of wire—with the question "What is this?" He would strike an arc, look into an eye-piece, and, within a minute or two, would tell you the main metallic elements in the sample; or, if it was a question of identifying a steel, he could tell you the main alloying elements. If you worked in a steelworks with a collection of reference samples, he could in a few minutes more identify the grade of steel, and perhaps its maker, by comparison with known standard samples.

Such analysts are rare—there may be four or five in Great Britain and perhaps a dozen in the U.S.S.R. (where visual spectroscopy is said to be more popular). This book may be reviewed by saying that it is indispensable to the training of such an analyst; the other qualification that he needs is two or more years of practice. The first edition of this book appeared in 1928, when it was one of the first handbooks for the spectroscopist, and this edition, completely revised in 1952, is more complete and more helpful. If you ever need it, its price will seem insignificant; but it is not a book for the general reader in metallurgy.

E. VAN SOMEREN